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WILFRED JOHN DAWSON, ASSOC.MET., F.I.M.

WILFRED JOHN DAWSON received his early education at Rotherham Grammar School and then became an articled pupil of Mr. F. J. Merrils, F.C.S., Sheffield, who had taken over the laboratory and consulting practice formerly conducted by Mr. J. O. Arnold. At the same time, he pursued the metallurgical course at the Technical School, Sheffield (now incorporated as the Applied Science Department of the University of Sheffield), under the late Professor J. O. Arnold, and at the age of eighteen was awarded the Associateship in Metallurgy.

After serving for three years in the laboratory of the works of Messrs. Henry Bessemer & Co., Ltd., he joined the laboratory staff of Messrs. Hadfields, Ltd., Hecla Works, Sheffield. Shortly afterwards he was put in charge of the steelmaking department of the Hecla Works and for nearly half a century he has been closely linked with all metallurgical developments of the firm.

He was closely associated with the late Sir Robert Hadfield, Bt., F.R.S., for a period of nearly forty years and assisted him in numerous researches which were undertaken during that period. During the 1914-1918 war he was promoted to the position of Manager of the Metallurgical Department, and joined the Board of Directors in 1919.

Mr. Dawson has taken a leading part in co-operative research, particularly during the last war, when he served as Chairman of the Technical Advisory Committee of the Special and Alloy Steels Committee (Steel Control of the Ministry of Supply), as a Member of the Metallurgy Committee of the Advisory Council on Scientific Research and Technical Development, and as a member of the Iron and Steel Technical Committee (Ministry of Supply). He was Chairman of the Tanks Cast Armour Committee from its foundation and a Member of the Tanks Wrought Armour and Aircraft Armour Committee, which contributed much to the development of the equipment of the fighting forces.

Mr. Dawson has served, or is now serving, as a member of more than twenty committees concerned with scientific research. Among these activities mention might be made of the following :

Member of the Engineering Divisional Council and the Iron and Steel Industry Committee of the British Standards Institution ; Chairman of the Steel Castings Research Committee, the Corrosion Committee, the Alloy Steels Committee, and the Hair-Line Crack Sub-Committee, formerly operated jointly by the Iron and Steel Institute and British Iron and Steel Federation. He has also acted as a Member of Council of the British Iron and Steel Research Association, and was a Member of the original Council of the Institution of Metallurgists, of which he is a Founder Fellow.

Mr. Dawson was awarded the E. J. Fox Gold Medal of the Institute of British Foundrymen, for 1940, in recognition of his work as Chairman of the Steel Castings Research Committee. At the Autumn General Meeting in 1947, he was elected an Honorary Vice-President of the Iron and Steel Institute.

W. J. Dawson, Assoc.Met., F.I.M.
Honorary Vice-President

IRON AND STEEL
INSTITUTE PAPERS

PLANT DESIGN · MANUFACTURE · RESEARCH · DEVELOPMENT

Construction and Repair of Open-Hearth Furnaces*

By D. C. Muir†

SYNOPSIS

An account is given of the furnace-repair programme at the works of the Consett Iron Co., Ltd., over a period of seven years and of a system of balanced repairs which enabled nine out of ten furnaces to be in production at any one time. Data regarding the construction, life, replacement, and costs of refractories in various parts of the furnaces are included, and proposals regarding the introduction of the all-basic furnace, furnace insulation, and the conversion of the furnaces from producer-gas to creosote-pitch, coke-oven-gas, or creosote-pitch/coke-oven-gas firing are discussed.

I—INTRODUCTION

THE output of steel from any particular furnace is controlled by a number of factors, but the most important is the availability of the furnace, *i.e.*, the percentage of time in which the furnace is capable of producing steel, or the ratio of the producing time to the rebuilding time. The steel-producing time, that is, the length of the furnace campaign, depends almost entirely upon four factors, namely, operation, design, construction, and refractories, all of which are interdependent.

There is little doubt that the pressure for increased production during recent years has resulted in decreased furnace life, and the experiences recorded later in this paper will show that it is very easy to increase that pressure past the optimum point and reach a state where overall production is reduced because the furnace life is shortened.

During the war years, design and construction remained almost static, because of (*a*) the urgent need for repairing the furnace as quickly as possible in order to maintain production, and (*b*) the uncertainty regarding the quantity and quality of the basic refractories available. Also, the choice of refractories has been limited owing to the lack of certain raw materials, and this limitation has had its effect on furnace life and refractories consumption. As it is now possible, however, to look forward to improved basic refractories for the construction of steel furnaces, and as the tendency towards the use of coke-oven gas and liquid fuels has permitted great variations in furnace design, the present seems a suitable time to review furnace performance during the past seven years and to examine the degree of efficiency achieved, and also to consider in which directions further improvements are necessary. It is the object of this paper to give some indication of the methods of construction and repair adopted in the melting shop of the Consett Iron Co., Ltd., in order to ensure that the maximum number of furnaces is in effective operation. The results of certain experiences of basic-brick construction, together with some opinions on the most promising developments in such installations and an account of proposals for increasing furnace efficiency, are also recorded.

II—THE FURNACES

(1) *General*

The melting shop consists of seven basic and three acid furnaces, all of the fixed type. The

* Received 6th March, 1947.
† Consett Iron Co., Ltd.

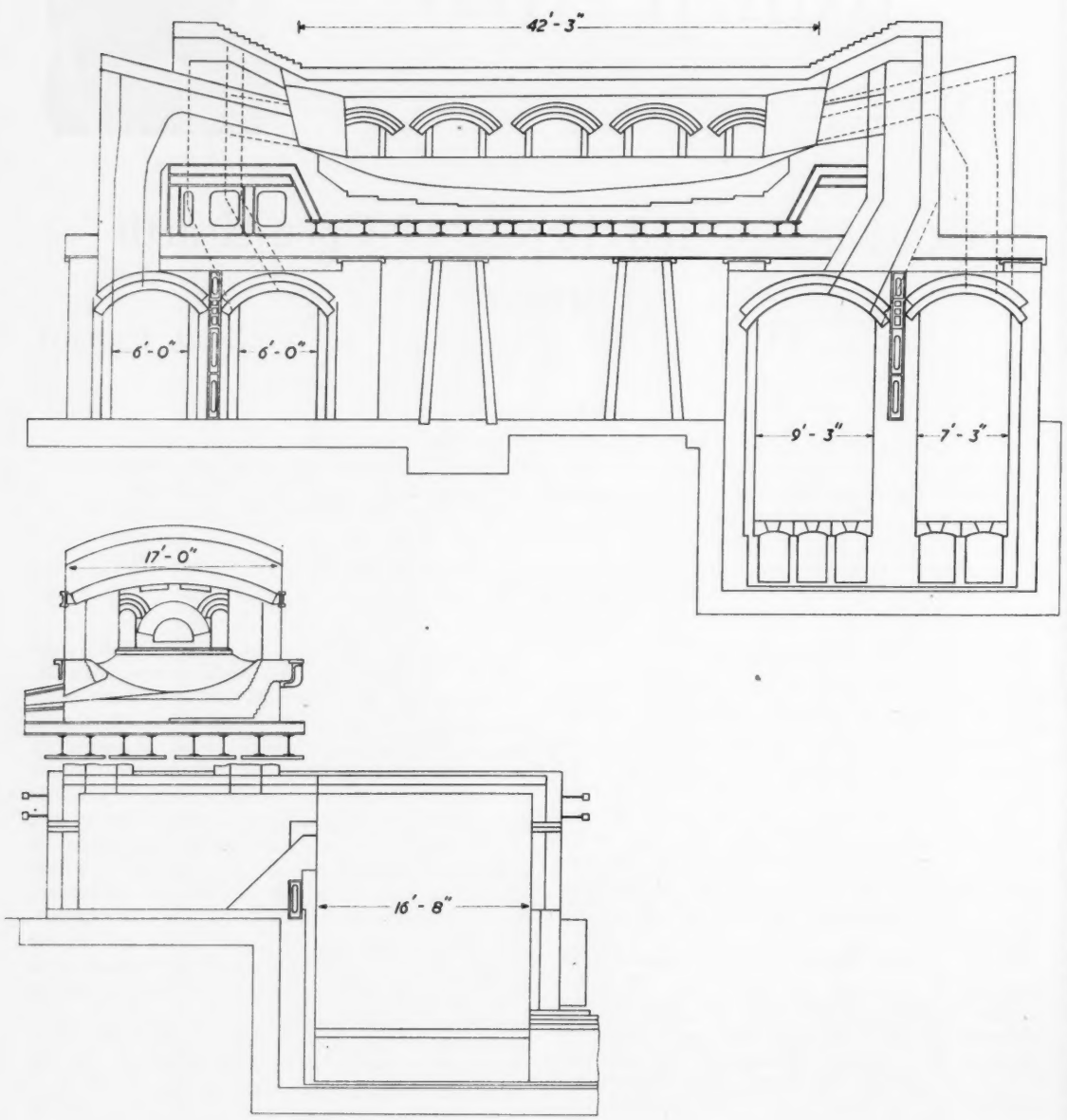


FIG. 1—Sections through 75-ton furnace

basic furnaces and two of the acid furnaces are of 75 tons nominal capacity, and the remaining acid furnace is of 40 tons nominal capacity. For the purpose of delineation only the 75-ton furnaces are described, but other data apply to all the furnaces, unless otherwise stated.

The furnaces were originally of Wellman design, but they have been altered in the course of years with regard to the size of air uptakes, slopes of the ramps and gas ports, and design of the gas and air ports, etc. There is no water-cooling of any kind. The fuel is raw producer gas with the addition of 5% of coke-oven gas (thermal basis), and the air is induced by fan or by the chimney. Details of the furnaces are given in Fig. 1, and a photograph is shown in Fig. 2.

The average weekly output per furnace during the period from January to July, 1946, was 967·16 tons for the basic furnaces and 872·06 tons for the 75-ton acid furnaces. The steel is made into ingots for the production of plates, billets, sections, rails, etc. The greater part of the basic steel goes to the plate mill. About 15% of the output of the acid furnaces consists of alloy steel, but during the war period this figure rose to as high as 45%.

(2) *Operation*

The furnaces were originally designed to operate on cold pig iron and scrap and did so until 1941, when it was found possible to use hot metal regularly for the basic furnaces. The average constitution of the charge since 1939 is shown in Table I.

After a rebuild the furnaces are heated-up with coke-oven gas introduced through the charging doors by means of three **T**-shaped pipes. Metal strips are placed on the roof and are measured against the cross-beams of the furnace structure, the strips being marked so that the rise and fall of the roof can be readily observed. After the Saturday-morning tap and the necessary repairs, the coke-oven-gas pipes are introduced and the doors sealed with clay. The furnaces have a lying-in charge and are gassed again at midnight on Saturday.

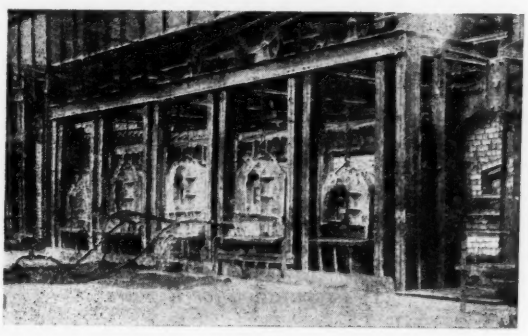


FIG. 2—Outside view of front of 75-ton furnace

TABLE I—*Principal Materials Charged Per Ton of Ingots Produced*

	Hot Metal, %	Cast-Iron Scrap and Pig Iron, %	Steel Scrap (bought and home), %	Lime and Limestone, %	Scale and Ore, %
Basic					
1939	...	29·52	61·05	7·38	2·05
1940	6·95	21·88	59·41	8·93	2·83
1941	22·73	9·72	52·65	11·04	3·86
1942	23·18	10·00	50·74	11·70	4·38
1943	20·34	9·10	56·52	10·36	3·68
1944	20·03	11·71	56·68	9·47	2·11
1945	22·56	6·26	60·25	9·65	1·28
Acid					
1939	...	40·16	55·62	0·25	3·97
1940	...	35·93	58·89	0·27	4·91
1941	...	27·11	69·38	0·19	3·32
1942	...	25·70	71·67	0·16	2·47
1943	...	25·37	72·54	0·24	1·85
1944	...	27·81	70·08	0·23	1·88
1945	...	30·66	67·14	0·16	2·04

(3) *Repairs*

Previous to 1940 the basic furnaces had been scheduled to run for 30 weeks without a major repair, and the acid furnaces for 24 weeks. The limiting factors to the life of the basic furnaces were the failure of the roof and the choking of the regenerators and flues, while the limiting factor in the acid-furnace campaigns was the filling-up of the slag pockets.

With the fall of France in 1940 the usual supplies of foreign ore were completely shut off, and only home ores were available in bulk. This change of material, together with the wide variations in the quality of scrap and the pressure for increased output, had its effect upon the lives of the roofs and regenerators of the basic furnaces, and it became necessary to shut furnaces down for replacement of these parts after 15–20 weeks. Such an alteration in practice required a rearrangement and an all-round speed-up of the rebuilding schedule. In the transition period before the establishment of the present system it was found that furnaces were being put off for major repairs to roofs, regenerators, etc., in such a way that it was not possible to ensure with certainty that one rebuild could be completed before the necessity for another major repair entailed the shut-down of a second furnace. The variation in roof life at this period is shown in Figs. 3 and 4.

III—THE SYSTEM OF BALANCED REPAIRS

Two parts of the furnace which necessitate a shut-down for repair work are the roof and the

checkers, and, in common with most steelworks, great variations in the lives of these parts of the furnace have been experienced owing to war-time conditions. With the object of ensuring that the maximum amount of the total furnace capacity could be kept in continuous production and in the most effective condition, a system of balanced repairs was instituted, by means of which the shut-down was so arranged that the maximum amount of repairs could be done during this period. The length of the acid-furnace campaigns was fixed at 24 weeks, as by that time the slag pockets were so full that they affected the draughting of the furnaces. The target for the basic furnaces was set at 30 weeks, with a shut-down about halfway for a top dressing. At the present time it is found that after 15 weeks the blocks, back lining, air uptakes, and roof are in a condition to benefit from a complete renewal, and at the same time the top ten courses of checkers are taken out and replaced.

The progressive approach to the target of maintaining nine furnaces in operation at any

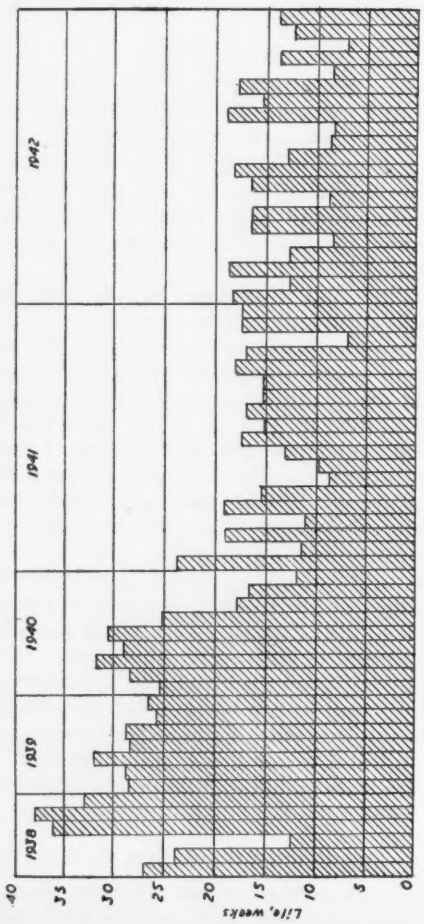


FIG. 3 (*a*)—Roof life, basic furnaces (*see also* Fig. 3 (*b*))

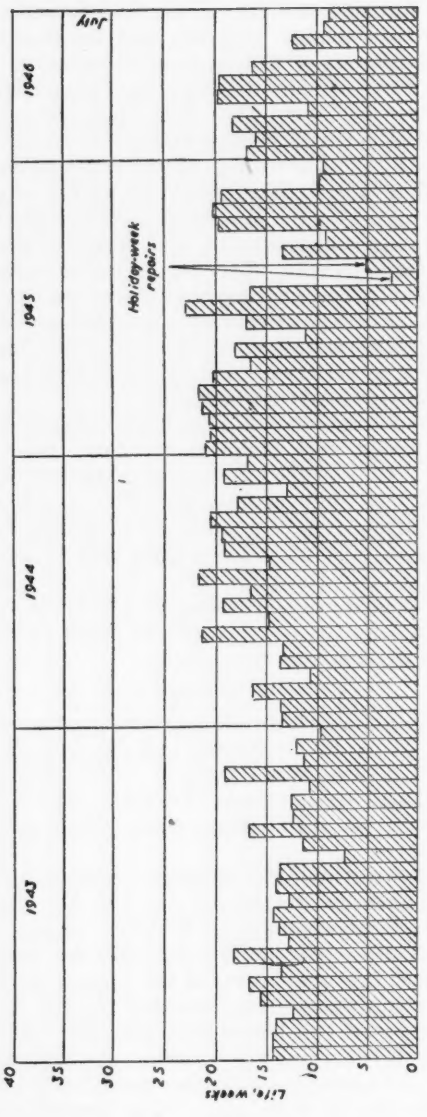


FIG. 3 (*b*)—Roof life, basic furnaces (*see also* Fig. 3 (*a*))

one time is indicated by the following data of furnace availability :

	Furnaces in Production, %
1939	85·8
1940	87·1
1941	85·8
1942	88·3
1943	89·5
1944	89·3
1945	90·3
1946†	91·4

† January to July

(1) *Preparation for Repairs*

(a) *General*

In demolition or constructional work there is

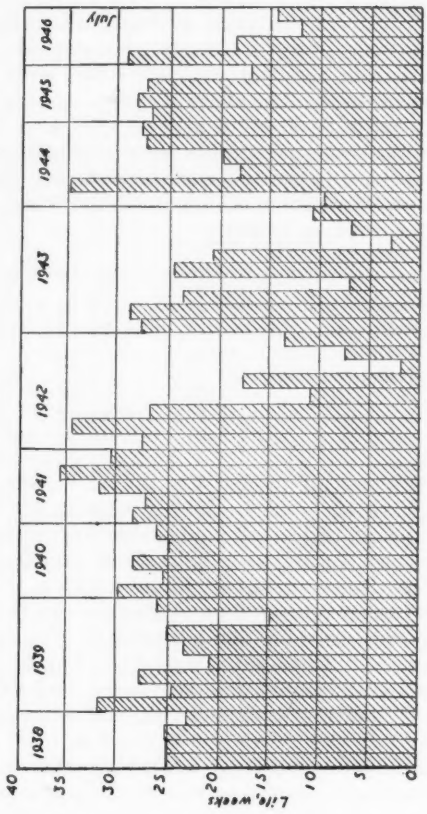


FIG. 4—Roof life, acid furnaces

an optimum number of bricklayers and labourers that can be efficiently used in conjunction with mechanical aid, and the problem is to organize the work in such a manner that the maximum number of men is employed for the shortest possible time. The arrangements for the removal of from 450 to 500 tons of scrap material from a demolished furnace and for the supply of building materials in their correct sequence on the site can also influence to a great extent the speed at which the furnace is rebuilt. Such arrangements require the closest co-operation with the melting-shop management and the traffic department, in order to avoid any congestion of traffic or interference with the melting-shop operations, and thought given to this aspect of the subject is well repaid in the results which can be achieved. The organization of the actual repair work depends on co-operation between the melting-shop management, the Refractories Department, and the engineering staff.

(b) *Consett Repair Practice*

The greater part of the refractory material used for furnace repairs at Consett comes from the Company's own brickworks, situated about a mile away from the melting shop. Arrangements are made to have the bulk of the material needed for a repair delivered in trucks as required, in order to avoid the double handling which would otherwise result from putting the material into stock in the melting shop and taking it out again.

To facilitate the handling of the bricks, specially designed metal containers with hook attachments are used. The containers are carried on flat wagons and can be lifted by the charging-bay crane directly on to the furnace platform, near the furnace to be repaired. An extension of the container system of handling, at present in the planning stage, will eliminate man-handling of the material almost completely.

The rebuilding of the furnace is normally carried out by three-shift working, and an endeavour is made to have as much as possible of the material required for each shift on the melting-shop platform before the shift commences; very close co-operation between the melting-shop staff and the bricklayer foreman is required to achieve this without causing any interruption to production. The distribution of the services of the cranes and charging machines between the constructional work and the normal operations must be carefully organized.

After the linings have been scrapped, the labourers start to dig around the hearth and banks to prepare for new linings, and a burner is always on hand to assist in cutting through any metal which may be present around the edge of the hearth. The scrapped material is loaded into charging trays and discharged into trucks *via* a chute on the melting-shop charging platform. If the furnace goes off on a Saturday, the reconstruction of the linings and roof is usually completed by the end of the first week. During this period a labour gang is engaged on the removal of slag from the slag pockets, while another is cleaning-out the flues and checkers. After the above-stage section of the furnace repairs is completed, most of the bricklayers can be transferred below stage, to repair the slag pockets and refill the checkers. While the repairs are in progress, a small portable cement mixer is kept on the charging platform, beside the furnace, to temper the cements and bring them to the proper condition for use on the spot. The checker bricks which are badly eroded or covered with slag are sent to the ballast heap, and the clean bricks are stacked for re-use. The recovery from the furnace of bricks which are capable of further use represents a considerable saving.

If any repairs to the constructional steelwork of the furnace are required, the materials are put in hand before the furnace is shut down, and the repairs are effected quickly, before the bricklaying starts. Except in the case of alterations to the

design of the furnace, there is normally little work for the fitters to do in this respect, apart from repairing the plates around the slag notch, tap-hole, and roof skewback channels. Occasionally the air coolers in the slag pockets require attention.

The average rebuilding time for a basic furnace, from the last tap to the first charge, is 2·4 weeks, the maximum being 2·73 weeks and the minimum 2·2.

(2) *Labour Force*

The maximum number of bricklayers employed is 50, and the general labourers and bricklayers' labourers total 100. There are also three charge-hands. From 3000 to 4000 bricklayers' man-hours and approximately 8000 labourers' man-hours are required.

The allocation of the labour force naturally varies from furnace to furnace, but the pulling-down of the furnace above stage level, disposal of the rubbish, and digging-out a foundation for the linings normally occupies a squad of from 20 to 25 labourers per shift for five shifts. Squads of from 10 to 15 bricklayers per shift, with attendant labourers, then move in, and in a period of nine shifts they complete the linings, roof, ramps, gas ports, and furnace ends. The general labourers have meanwhile moved downstairs and started upon the cleaning-out of the slag pockets and checker chambers. There is no regulation manner of applying the labour gang below stage, as the work must be arranged so that there is no hold-up of the bricklayers who follow on. For example, if repairs to a slag chamber or a regenerator roof are necessary, then as much labour as possible is centred on the particular job, in order to prepare it for the bricklayers, before moving on to the other work.

For all the work below stage an average of 16 labourers per shift is employed for 30 shifts. Of this number a special gang is usually entrusted with the cleaning-out of the slag chambers, and this job occupies an average of six men per shift for nine shifts. For the rebuilding of the slag-chamber walls, the uptakes, and the regenerator-chamber roofs, together with the filling-up of the checker chambers and the closing-up of the furnace, 21 shifts of from 10 to 15 bricklayers (plus attendant labourers) per shift are required.

The net rebuilding time varies between 12 and 15 days, the principal factor influencing the time being the amount of repair work required on the slag chamber and the regenerator arches.

IV—Furnace Reconstruction and Maintenance

(1) *Roofs*

The roofs of the Consett furnaces are built of $12 \times 6 \times 3$–$2\frac{7}{8}$-in. end-arch silica bricks in a bonded construction, without cement, and are curved only from the front to the back lining, being flat from ramp to ramp. No expansion space is left in the roof, which is built with a rise of $1\frac{1}{4}$ in. to the foot of the chord.

As has already been mentioned, previous to 1940 the basic-furnace roofs were expected to last 30 weeks, and the acid-furnace roofs 24. The average roof-lives for the basic and acid furnaces since 1939 were as follows:

	Basic Furnaces, weeks	Acid Furnaces, weeks
1939	28·48	24·32
1940	22·73	21·94
1941	15·25	30·65
1942	13·65	16·77
1943	13·63	17·02
1944	16·73	22·89
1945	15·91	24·26

For purposes of record, the life of a roof is considered to be ended when a large patch is required, as it is deemed inadvisable to compare the lives of patched roofs with those that have completed their campaigns without a patch.

After 1940, when supplies of foreign ores were restricted and iron for the manufacture of basic steel had to be produced mainly from lean home ores, the extra refining time very rapidly affected the roof life of the basic furnaces, as is shown in the data just given and in Fig. 3. The roofs of the acid furnaces were not affected to the same degree, but towards the end of 1942 there was a serious reduction in the life of both acid and basic furnace roofs, owing to the pressure for increased output and a considerable increase in the proportion of special steels produced. It will be noticed that several roofs had very short lives indeed, and the rebuilding situation became serious.

Some years before the war it had been proved from conscientious observation and a study of furnace records that careful glazing of the roof and the avoidance of any burning during the first few weeks ensured that the roof would have a reasonable life. The beneficial effects of glazing the roof have since been accorded a scientific basis by the work of the Open-Hearth Refractories Joint Panel in their report* on the durability of basic open-hearth furnace roofs, which showed, *inter alia*, that the conductivity of a roof after glazing for a few weeks was considerably higher than that of an unglazed silica roof.

When pushing for output, new furnaces were sometimes called on for a cast of special steel before the roof was glazed. The effect of this treatment was quickly shown by premature burning of the roofs and decreased roof life. The reduction

* *Journal of the Iron and Steel Institute*, 1941, No. II, pp. 203P–262P.

in life was so serious that instructions were given that no furnace should be put on special steels until it had been in operation for at least two weeks, and that all new roofs should receive careful treatment. The immediate effect of these instructions in promoting a steady increase in roof life is shown in Fig. 3. The life of a normal basic roof now varies between 14 and 20 weeks.

(2) *Slag Pockets*

The slag pockets are emptied only at the end of a campaign. Formerly the slag was then from 4 to 5 ft. high in the pocket and consisted of a solid vitrified mass firmly fixed on the pocket walls. The only way to remove this was to dig into the silica wall of the slag chamber and remove the slag in very large pieces with the aid of steel cables and a 100-ton crane. This work was very hot, arduous, dirty, and slow, and the labourers were paid a special rate for doing it. An attempt to hasten this work by the use of explosives was made but was not successful, as this method was found to be just as costly and no quicker than the old method, for the following reasons: (1) Owing to the height of the slag in the slag pocket, it was not possible to drill holes vertically over the surface of the slag, and they had to be put in from the front; (2) the slag was so hard that the best of steel drills did not last long and made slow progress; (3) when the explosion took place the material was inclined to blow out into the casting bay, and time-occupying precautions had to be taken to prevent this.

Several other means of getting rid of the slag were tried without success, but the method which was eventually adopted consists of building a 9-in. false wall inside the slag chamber and packing the 3-in. space between the false wall and the slag-chamber wall proper with coke breeze. This construction has been found to be the most successful form of resistance to slag attack and permits a quicker removal of the slag, as the coke can be more readily removed than the brickwork, or a space is left which forms a starting point for further removal of the slag.

About 4 in. of a chrome ramming compound are used on the bottom of the slag chamber in order to prevent penetration of the slag into the brickwork and foundations of the furnace.

(3) *Hearths*

For many years the basic-furnace sub-hearths were built with 70/30 magnesite-chrome brick, except for a quantity of magnesite around the tap-hole, but for the past few years they have been constructed entirely of stabilized dolomite bricks. In the course of the last three years a number of hearths have been made by ramming graded dolomite and tar, and these have proved quite satisfactory. The greatest improvement, however, has been in the maintenance of the hearths and the formation of the tap-holes with special mixtures based on stabilized dolomite, which have enabled good tap-holes to be made and rapid repairs to the hearth to be effected. The improvement resulting from the use of these materials is indicated in Table II, which shows the decrease in the dolomite consumption and in the time lost owing to hearth trouble in the basic furnaces during the period mentioned.

TABLE II—*Results of Use of Special Mixtures Based on Stabilized Dolomite*

	Dolomite Consumption		Working Time Lost, %
	Tons	Lb./Ton of Steel	
1943	9914	77·14	5·07
1944	9448	74·43	4·27
1945	8737	68·39	3·25

A new system of records has been introduced at Consett whereby it is hoped to establish some definite correlation between the life of the basic hearths and the dolomite consumption and, in time, to determine the economic life of a hearth.

Any holes which occur in the furnace hearth or banks, the position of such holes, and the extra time taken to fettle them, are marked on a special card which shows a plan of the hearth, together with a record of the dolomite consumption for the entire furnace. From this card it will thus be possible to obtain a very good idea of the condition of any hearth during the course of the furnace campaign.

(4) *Checkers*

(a) *Basic-Furnace Checkers*

Previous to 1939, fireclay was used for the bottom portion of the basic-furnace checkers and silica for the top 12 courses. The checkers lasted throughout the furnace campaign of about 30 weeks, but by the end of that time the top 2 ft. had either collapsed or had disappeared to form a thin slag, which ran down to a lower portion of the checkers but did not tend to fill up the openings between them. In 1940, when the roof life began to decrease owing to the change in practice, the extra lime and iron oxide also had their effect upon the checkers, and it was found necessary to renew the first ten courses after about 15 weeks. In fact, the failure of the regenerators kept pace with the failure of the roofs.

High-alumina firebricks were then tried in place of the silica bricks. The alumina bricks resisted chemical attack by the slag, but it piled up on

TABLE III—*Analyses of Flue Dusts from below Checker Bearers*

	N.E. Air-Checker Bottom	N.E. Gas-Checker Bottom	S.E. Air-Checker Bottom	S.E. Gas-Checker Bottom
Silica, %	2·50	18·10	2·70	15·50
Ferric oxide, % ...	39·98	39·68	41·70	35·38
Alumina, % ...	1·71	11·46	1·63	10·26
Titanic oxide, % ...	0·12	0·52	0·10	0·46
Chromic oxide, %...	0·17	0·45	0·19	0·51
Vanadium pentoxide, %	Nil	Trace	Nil	Trace
Phosphorous pentoxide	0·72	1·10	0·77	0·92
Manganous oxide, %	0·69	0·63	0·88	0·68
Lime, %	6·30	8·20	5·90	7·60
Magnesia, % ...	1·86	2·43	1·44	2·64
Zinc oxide, % ...	12·95	10·00	13·40	14·10
Lead oxide, % ...	10·50	0·18	8·10	0·18
Copper oxide, % ...	0·27	0·30	0·30	0·25
Stannic oxide, % ...	0·20	0·10	0·20	0·10
Sulphur trioxide, %	19·25	4·40	20·50	8·55
Alkalies, % ...	2·00	1·74	1·92	1·80

top of the first few courses of the checkers and impeded the draught. After more experiments semi-silica bricks of high porosity were tried and were found to be very successful on the top of the checkers, as they have the ability to absorb the slag to some depth without melting or becoming distorted.

The analyses of some flue dusts taken from below the checker bearers are listed in Table III and give some idea of the slag attack on the regenerator brickwork.

(b) *Acid-Furnace Checkers*

No difficulty is experienced with the checkers of the acid furnaces. The gas regenerators are filled in a similar manner to those of the basic furnaces, that is, with 9 × 4½ × 3-in. fireclay bricks topped with 9 × 4½ × 3-in. siliceous bricks. The bricks become covered with a thick viscous slag during the course of the campaign and have to be renewed.

The air regenerators are built with 12 × 6 × 3-in. fireclay and silica bricks, set as box checkers. Slots are left in the regenerator-chamber arch so that the box checkers can be rodded down and cleaned out at intervals, normally every two weeks, during the campaign. These air checkers last for many years and are renewed only when the bricks are broken in the course of handling, as they tend to become fragile after several campaigns.

(5) *Basic-Furnace Back Walls*

Previous to 1939, extensive trials were carried out in which various types of brick were used on the all-basic back wall; the last such trial was made with a lining constructed of 14 × 6 × 3-in. Radex *E* bricks. The maximum life obtained with the Radex bricks on any of the furnaces was 11 weeks, which was little better than that obtained when using the home-produced chrome-magnesite bricks. When supplies of magnesite and chrome-magnesite were cut down early in the war, the old type of construction, in which the top half of the wall is built with silica brick, was reverted to. Attempts were made to increase the amount of silica brick in the back wall, but it was found that this actually had the effect of increasing the consumption of chrome-magnesite bricks, as the flame cut into the lower part of the wall and the resulting slag ran down on to the chrome-magnesite bricks below, with the result that the back walls had to be replaced every three to four weeks. In order to repair the wall, a portion of the chrome-magnesite brickwork adjacent to the silica bricks had to be taken out, and the amount of chrome-magnesite consumed during these frequent repairs was greater than that used when the chrome-magnesite was taken more than halfway up the back wall, where it, and not the silica brick, was subjected to the direct contact of the flame.

During the last two years the back walls of the basic furnaces have been converted, whenever possible, from the straight to the semi-sloping type, and these alterations have resulted in increases in the lives of the back linings. The average life with a straight back wall is approximately 9 weeks, whilst an average life of 16 weeks is attained with the semi-sloping type. Very good results have been obtained with chemically bonded chrome-magnesite built "on the batter" with 14-gauge metal plates on the bedding face, and this construction is now standard.

When the back lining becomes thin it has a tendency to bulge. To prevent this and to give additional support to the lining, light steel strips or angles are fixed between the buckstaves.

As the hot metal is introduced through a door in the back wall, it has not been possible to build a fully sloping back wall, as this would interfere with operations in the casting bay.

(6) *Uptakes*

When rapid rebuilding of the furnaces became a matter of importance, it was found that the top portion was very often completed before the slag was out of the slag chambers. Until this slag was cleared and the bonnet of the slag chamber was complete, it was not possible to start work on the uptakes and the ends of the furnace above stage level; there was thus a delay before the bricklayers could resume work, and this resulted in an extension of the rebuilding period. Some

angles on the buckstaves were therefore introduced, just below stage level, and a relieving arch was built across the uptakes. This permitted the construction of the above-stage end walls of the furnace to be proceeded with, even if the repairs to the slag-chamber arches were not then complete.

The uptakes are constructed in silica, and very serious erosion is found at the air and gas ends. A number of trials of the use of chrome and chrome-magnesite bricks as a lining for the above-stage portion of the air uptakes were therefore carried out. These trials proved very successful, both in the reduction in the cost of refractories and in the man-hours required for repairs ; details are given in Table IV.

It is now standard practice to line the air uptakes with basic brick, and no repairs are necessary at the air ends until the furnace is top dressed. Trouble has been experienced in the gas uptakes owing to erosion of the silica in the lower portion, and it is therefore proposed to construct these uptakes in chrome-magnesite brick from the slag-chamber bonnet in order to overcome this trouble. An indirect result of this basic construction in the uptakes has been a very considerable reduction in the quantity of slag in the air slag chamber.

(7) *Gas Ports*

Special silica blocks are used in the construction of the gas ports, as it is found that they give a longer service than rings of 9-in. bricks. The disadvantage of the silica gas port is that it wears back and in effect increases the length of the furnace, while the inside of the port wears away, thus increasing the port area and reducing the gas velocity, and increasing the tendency of the flame to spread to the linings and to lift on to the roof. The original blocks have a life of from 9 to 10 weeks, after which they are repaired by Scotch-blocking. The Scotch block lasts a further 5–6 weeks, by which time the furnace is usually ready for the top dressing. The building of a good Scotch block depends to a great extent upon the skill of the bricklayers.

TABLE IV—*Comparison of Costs and Working Time for Chrome-Magnesite and Silica Linings*

	Cost per Furnace per Week	
	Chrome-Magnesite	Silica
Materials*	£6 10 0	£6 10 6
Labour	£0 9 7	£2 4 0
Total	£6 19 7	£8 14 6
Man-hours	4·08	19·02

*Including cost of initial installation

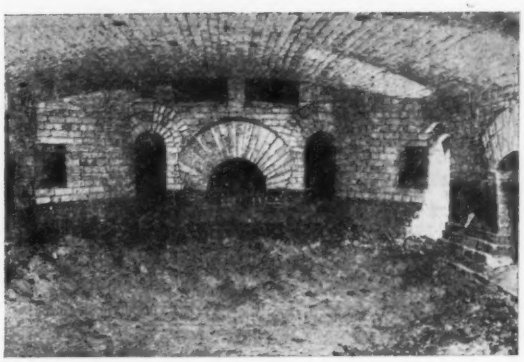


FIG. 5—Gas and air ports

An effort made to build the gas ports in chrome-magnesite brick was not successful, as the present design of the gas port and the side air ports is not sufficiently stable for a basic-brick structure, which lacks the mechanical strength of a silica-brick structure, in which the bricks become fluxed together.

The gas and air ports are illustrated in Fig. 5.

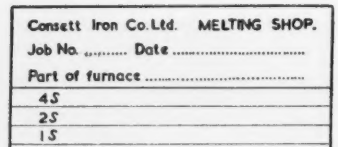
(8) *Front Lining*

The front lining is of orthodox construction, consisting of chemically bonded chrome-magnesite brick up to the spring of the arch, with silica brick for the remainder. As there is no water-cooling and a large proportion of scrap is charged, there is a fair amount of mechanical damage, and repairs to the door jambs and arches are comparatively heavy. The jambs are patched with high-porosity silica bricks which have been soaked in tar, and these are found to be much better than the normal silica bricks. From the point of view of refractories costs and the prevention of cold-air ingress, it is considered that the doors and door jambs are the only parts of the furnace where water-cooling would be justified.

V—REFRACTORIES : RECORDS AND TESTING

(1) *Records*

Records of all refractories stocks, consumptions, and costs are kept by the Refractories Department in the steelworks. The records of the consumption are based on a system of chits which are issued to the foreman bricklayer and by him to the bricklayers and labourers. All the normal bricklaying materials are itemized on these slips, and the labourers have only to set down the quantities against each item and return the slips to the foreman, who forwards them to the Refractories Department, where the figures are entered up



Consett Iron Co. Ltd. MELTING SHOP.
Job No. ..... Date .....
Part of furnace .....

4S
2S
1S
1C
1FS
1FE
1
1FS (Tars)
1 (Tars)
1B
4B
21SS
27B
4B (Tars)
21SS (Tars)
9"x4½"x3" Chrome-Mag.
Silica Mud
Chrome Paste
Number of Bricklayers
Number of Labourers
Time started Time finished
Name.....

FIG. 6—Type of slip issued for Saturday repairs

and detailed. Records are therefore available for the consumption of refractories, the amount of labour used, and the construction and maintenance costs for each part of the furnace. The type of slip issued for Saturday repairs is shown in Fig. 6.

The refractories consumption per ton of steel for basic and acid furnaces is as follows :

	Consumption, lb./ton of steel	
	Basic Furnaces	Acid Furnaces
Silica brick	37·7	45·16
Chrome-magnesite brick	5·75	...
Magnesite brick ...	...	...
Stabilized dolomite brick	1·74	...
Fireclay brick	8·80	7·22
Siliceous brick	5·36	1·86
Silica clay	13·8	24·20
Basic cements	2·83	...
Fireclay	0·22	0·30
Refractory chrome ore	1·86	...

(2) *Testing*

All the silica and basic bricks and a large proportion of the firebricks used in the works come from the Company's own brickworks, and test data are therefore readily available. The inbrought materials are regularly sampled and tested in the brickworks' laboratory. The co-ordination of laboratory tests and works trials, and the observation and recording of the behaviour of refractory materials in service, are the responsibility of the Refractories Department in the steelworks.

VI—FURNACE DEVELOPMENT

Ideas for the development of the Consett furnaces can be set down under three principal headings, namely, (1) increased availability, (2) increased rate of production, and (3) reduced operating costs.

(1) *Increased Availability*

There are two methods of increasing the availability of the furnace, *viz.*, (*a*) by reduction of the number and length of the shut-down periods and (*b*) by reduction of the time taken for rebuilds and repairs. The reasons for shutting-down the furnaces are usually the failure of the roof, the blocking-up of the checkers and flues, and the filling-up of the slag pockets.

(a) *Reduction of Shut-Down Periods*

The All-Basic Furnace—The all-basic furnace seems to afford the greatest promise of any appreciable extension in the roof life, but the scanty records available for all-basic furnaces working under the conditions generally prevailing at Consett present insufficient data for the forming of any definite judgment in the matter. The results obtained on the Continent and in America, under conditions of a 168-hr. working week, high percentages of hot metal, the duplex process, etc., cannot be taken as a sure guide to the results which are likely to be obtained on a furnace charging 60% of steel scrap and subject to week-end shut-downs.

The week-end shut-down, which is common practice in this country,* cannot have a beneficial effect on a basic roof, even with furnaces for which coke-oven gas is available for the maintenance of a red heat. Nevertheless, there is no doubt that in time basic roofs capable of lasting a year without repair will be available. Even if they should last only 30 weeks, *i.e.*, twice the length of the normal silica-roof life, the extra cost would be justified. However, the development of a suitable roof for each type of furnace will certainly be a slow and difficult process and offers plenty of scope for the ingenuity of plant engineers, bricklayers, and brickmakers.

The difficulty we foresee with the all-basic furnace with our present design is that of increasing the lives of the slag pockets, checkers, and flues to something approaching the life which might be expected from the furnace roof, so that the balanced-repair programme, together with the increase in the length of the furnace campaign, can be maintained.

Checkers—We hear with envy of checkers in America and Germany which last for years without replacement, but in Consett practice, which entails the charging of 60% of mixed scrap, we find that the checkers of the basic furnaces are blocked-up with slag in from 15 to 20 weeks,

* Agreements have since been made whereby in some districts the week-end shut-down has been reduced or eliminated.

in sufficient degree to affect the efficiency of the furnace, and we know of no immediate answer to this problem. The slag does not settle in the bricks in the form of dust which can be blown off by a steam or air jet, but as a heavy slag which adheres to the brickwork to such a degree that hammering is necessary to separate the bricks during dismantling. As it is not possible to prevent the slag adhering to the checkers, the logical step would appear to be a reduction in the velocity of the waste gases so that the greater part of the slag is deposited in the slag chamber. This may be done by increasing the area of the uptakes or the volume of the slag pocket, but there is some doubt if such methods can, within the limits of the existing furnace design, reduce the amount of slag in the checkers to such an extent as to lead to a checker life equal to that to be expected from a basic roof. A curtain wall across the centre of the slag chamber has been tried in an endeavour to accelerate the deposition of slag, but the effect was negligible ; also, when 2 or 3 ft. of slag had accumulated in the slag pocket, the curtain wall influenced the draughting of the furnace and had to be knocked down.

There is reason to believe that the conversion of the furnaces from raw producer-gas firing to either coke-oven-gas or oil-fuel firing will result in greatly increased checker life, as it will then be possible to increase the volume of the slag chamber by combining the existing gas and air slag chambers, and thus reduce the velocity of the waste gases. The construction of the uptakes in basic bricks will greatly reduce the quantity of slag in the slag pockets, and a twelve-months campaign can be expected from this section of the furnace.

Flues—At Consett the flues contain about 1 ft. of dust after a campaign of 30 weeks, and this dust is removed by bucket, through a manhole. This flue deposit is, however, of a friable and sandy nature, and it could be removed by high-pressure steam or air or by one of the systems now available for the purpose, so that clean flues could be maintained throughout the furnace campaign.

(b) *Reduction in Rebuilding and Repair Time*

The speed of the rebuild, of course, also directly affects the availability of the furnace, and any saving of the rebuilding time is therefore important and highly desirable. At Consett the friable nature of the slag in the slag pockets will permit speeding-up of repairs to that part of the furnace, and the extension of mechanical handling devices for the supply of material to the bricklayers and for the removal of refuse from the site should enable a 75-ton furnace to be rebuilt in less than two weeks from tap to tap. In brief, the aim is to construct a furnace which will have a minimum campaign of 30 weeks and a probable one of 50 weeks without any major repairs, and which will be rebuilt within a period of two weeks.

(2) *Increased Rate of Production*

Whatever the difficulties of basic roofs from a constructional point of view, there is no doubt as to the benefits to be derived from them in operation. The silica roof is a factor limiting the furnace temperature and, through that, the heat input and consequently the steel output. The basic roof is therefore desirable as an aid to increased output, and it is considered that the full benefits obtainable from the use of high-quality fuels will be realized only when such fuels are utilized in the all-basic furnace.

(3) *Reduced Operating Costs*

The reduction in the shut-down period which must arise from a longer campaign, and the increased production rate which would be made possible by the application of rich fuels to the all-basic furnace, must result in a greater output of steel and therefore in reduced operating costs.

It is also the intention to insulate the above-stage section of the furnace and so conserve fuel, which seems likely to become an increasingly costly item in furnace operation. The degree to which insulation can be applied will depend upon the quality of the refractories available and the construction of the linings and the roof; for example, upon whether the roof is sprung, semi-suspended, or fully suspended. The ultimate target is the development of a hot-face refractory insulating brick, and works tests are now in hand with high-porosity bricks of basic materials.

(4) *General*

To sum up, it is the intention at Consett to convert the furnaces to liquid-fuel and/or coke-oven-gas firing with properly located and efficient burners, to construct the furnace with basic material in order to ensure a long campaign, and to insulate the furnace structure in order to reduce heat losses.

Since the writing of this paper, one of the 75-ton basic furnaces has been pulled down and a 150-ton basic furnace with enlarged slag pockets, regenerators, and flues has been constructed on new foundations. This was done in less than nine weeks.

VII—Acknowledgment

The author thanks Mr. H. Boot, General Manager of the Consett Iron Co., Ltd., for permission to publish this paper.

Some Properties of Titanium Steels*

By L. Northcott, D.Sc.† and D. McLean, B.Sc.‡

SYNOPSIS

The effect of up to 6% of titanium on the structure and properties of plain carbon steel, containing 0·1–1·0% of carbon, and of four low-alloy steels (chromium–molybdenum, manganese–molybdenum, chromium, and nickel) has been investigated. Vertical sections showing the constitution of the Fe–Ti–C system at constant titanium contents have been prepared from the results of micro-examination, hardness tests, and thermal analysis. Tensile tests at room and at elevated temperatures, in conjunction with hardness tests on quenched and tempered specimens, show that: (i) When the titanium/carbon ratio is 4 : 1 by weight (1 : 1 atomic), high strength and hardness after quenching are maintained up to 600° C., provided a high quenching temperature—e.g., 1235° C.—is employed to ensure the solution of titanium carbide; 1% of chromium accentuates this effect; and (ii) when the titanium content is 4% or more, precipitation-hardening by Fe_2Ti occurs at temperatures in the region of 500°–700° C.; 3% of nickel accentuates the hardness increment and increases the rate of hardening. The results of other workers on the elimination of quench and strain-age hardening by titanium have been to a great extent confirmed. Titanium increases hardenability (Jominy test) to about the same extent as chromium, if a high quenching temperature—e.g., 1235° C.—is employed. Alloying elements increase the depth of hardening approximately in proportion to their own effect of hardenability.

INTRODUCTION

A RECENT review[1] of published information upon titanium and its effect upon iron and steel indicated that, although small additions of titanium are used for particular applications, information about its influence on the general mechanical properties of steel and about the iron–titanium and iron–carbon–titanium systems was somewhat conflicting and incomplete. A fairly comprehensive investigation was therefore commenced, covering in the first instance the influence of titanium on the properties of plain carbon and some low-alloy steels. The work was commenced before the war but was interrupted before much had been done on alloy steels. The work on plain carbon titanium steels was nearly completed, however, and is reported in Section I; results obtained with the low-alloy steels are given in Section II.

I—TITANIUM IN CARBON STEELS

Six series of steels were obtained, varying in carbon content from about 0·1 to 1%, the titanium content of each series varying in eight steps from 0 to 6%; there were thus 48 compositions in all.

The constitution of the steels was investigated by microscopical examination of, and hardness tests on, quenched samples, and by thermal and dilatometer curves; the mechanical properties of furnace-cooled specimens were determined, and the influence on hardness of tempering after quenching was investigated. Some corrosion tests were also performed.

Preparation of Material

The steels were prepared by the United Steel Companies, Ltd., in a 15–20-lb. spark-gap high-frequency furnace. On account of the commerical bearing of this part of the investigation, a special atmosphere was not employed, despite the well-known pronounced readiness of titanium to combine with nitrogen. Titanium additions were made with ferrotitanium for alloys of low titanium content and with titanium metal for those of higher titanium content; the analyses of these additions are given at the foot of Table I. The charge was melted in magnesite crucibles, free from silica, cast into an iron mould, and forged to $\frac{5}{8}$-in. dia. bar. No undue difficulty in forging was experienced. Forging was carried out on the basis of carbon content less one-fourth of the titanium content, *i.e.*, the temperatures used were somewhat higher than would normally have been used for titanium-free steels of the same carbon contents. The 4% and 6% titanium steels forged satisfactorily within the range 1150°–1050° C.

On receipt all bars were annealed for $\frac{1}{2}$ hr. at 900° C. and furnace-cooled before sectioning.

* Received 31st July, 1947.

† Ministry of Supply, Armament Research Department, Woolwich.

‡ National Physical Laboratory, Teddington.

TABLE I—*Chemical Composition of Titanium–Carbon Steels*

(Specimens marked *X* were repeat casts)

Specimen	C, %	Ti, %	Mn, %	Si, %	P, %	S, %	Al, %
A 0·6	0·11	Nil	0·52	0·12	0·01	0·025	<0·005
A 0·1	0·11	0·08	0·39	0·11	0·01	0·02	0·023
A 0·25	0·09	0·24	0·69	0·22	0·01	0·01	0·052
A 0·5	0·08	0·50	0·59	0·11	0·01	<0·01	0·09
A 1·0	0·11	1·40	0·53	0·16	0·015	0·01	0·26
A 2·0	0·10	2·20	0·56	0·06	0·01	0·01	0·52
A 4·0	0·09	4·02	0·67	0·09	0·02	0·02	1·02
A 6·0	0·08	6·08	0·55	0·09	0·025	0·02	0·85
A 6·0*X*	0·05	5·02	0·78	0·22	<0·01	<0·01	0·91
B 0·0	0·21	Nil	0·53	0·23	0·01	0·015	...
B 0·1	0·25	0·10	0·68	0·21	0·01	0·02	...
B 0·25	0·23	0·32	0·61	0·11	0·01	0·01	...
B 0·5	0·23	0·64	0·62	0·12	0·01	0·01	...
B 1·0	0·25	1·05	0·56	0·20	0·015	0·01	...
B 1·0*X*	0·23	0·85	0·67	0·13	0·015	0·025	...
B 2·0	0·27	2·25	0·66	0·08	0·02	0·01	...
B 4·0	0·24	4·96	0·69	0·09	0·025	0·025	1·36
B 6·0	0·22	6·12	0·70	0·21	0·025	0·025	1·09
C 0·0	0·38	Nil	0·68	0·22	0·01	0·025	...
C 0·1	0·39	0·08	0·64	0·18	0·01	0·015	...
C 0·25	0·38	0·24	0·62	0·11	0·01	0·01	...
C 0·5	0·37	0·68	0·61	0·14	0·01	<0·01	...
C 1·0	0·35	1·00	0·54	0·21	0·02	0·01	...
C 2·0	0·41	2·30	0·57	0·14	0·01	0·01	...
C 4·0	0·41	4·25	0·74	0·10	0·025	0·03	1·41
C 6·0	0·37	5·36	0·74	0·14	0·025	0·03	1·48
D 0·0	0·52	Nil	0·61	0·27	0·01	0·025	...
D 0·1	0·56	0·06	0·55	0·22	0·015	0·02	...
D 0·25	0·54	0·27	0·64	0·11	0·01	<0·01	...
D 0·5	0·50	0·72	0·64	0·13	0·015	0·015	...
D 1·0	0·53	1·10	0·63	0·14	0·015	0·01	...
D 2·0	0·52	2·30	0·65	0·09	0·02	0·015	...
D 4·0	0·50	4·60	0·74	0·11	0·025	0·025	1·02
D 6·0	0·60	5·83	0·73	0·18	0·025	0·02	1·15
E 0·0	0·81	Nil	0·58	0·23	0·01	0·02	...
E 0·1	0·76	0·10	0·60	0·16	0·01	0·01	...
E 0·25	0·74	0·26	0·62	0·17	0·01	<0·01	...
E 0·5	0·72	0·60	0·56	0·20	0·01	<0·01	...
E 1·0	0·74	0·75	0·63	0·11	0·015	0·015	...
E 2·0	0·77	2·70	0·66	0·08	0·015	0·01	...
E 4·0	0·67	4·34	0·72	0·14	0·025	0·025	1·51
E 6·0	0·83	5·48	0·75	0·12	0·02	0·02	1·24
F 0·0	1·10	Nil	0·57	0·29	0·015	0·02	Nil
F 0·1	1·16	0·085	0·55	0·27	0·015	0·02	0·019
F 0·25	1·07	0·26	0·52	0·21	0·015	<0·01	0·050
F 0·5	1·06	0·62	0·64	0·15	0·015	<0·01	0·10
F 1·0	1·00	1·10	0·65	0·13	0·02	0·01	0·26
F 2·0	1·05	2·20	0·67	0·09	0·02	0·01	0·51
F 4·0	1·03	4·70	0·73	0·12	0·025	0·015	1·07
F 6·0	1·09	5·65	0·72	0·11	...	...	...
F 6·0*X*	1·13	6·06	0·79	0·16	0·03	0·025	1·26
					Fe, %		
Ferrotitanium	0·07	41·7	...	0·79	38·3	...	16·4
Titanium metal	0·28	90·3	...	1·11	5·3	...	0·35

Chemical Composition

The analyses of the bars as received are given in Table I. The numbering of the alloys follows an obvious scheme. It will be seen that the carbon contents are fairly uniform in each group and that the titanium contents increase regularly for each carbon content. The other elements normally determined are reasonably low or constant. Manganese in nearly all cases is between 0·55 and 0·75%, the most serious exception being *A* 0·1, containing 0·39%. Silicon is rarely above 0·25%, the most serious exception in this case being *A* 6·0*X*. Phosphorus does not exceed 0·025%, but sulphur does so in two instances. The aluminium contents, however, are seriously high, being about one-quarter of the titanium content. The aluminium was presumably introduced *via* the titanium addition and it seems either that the sample of titanium metal available for analysis had an exceptionally low aluminium content or that additions were made by ferrotitanium throughout.

The steels are therefore in effect titanium–aluminium steels, and this has been borne in mind when interpreting the results. Several effects due to the alloy additions are described below. In most cases the decision as to whether an effect was caused by titanium (with aluminium doubtless playing a secondary rôle) was made according to whether or not it is governed by the titanium/carbon ratio. The effect of aluminium presumably depends either directly on the aluminium content or on the aluminium/oxygen ratio, and neither of these follows the same course for varying compositions as the titanium/carbon ratio.

Constitution

Microscopical examination revealed some constituents other than the usual steel phases :

(*a*) The well-known titanium nitride–carbide type of inclusion ; yellow, angular particles of titanium nitride (TiN) and grey angular particles of titanium carbide (TiC) were found, with a fairly continuous gradation yellow – brown - yellow–brown-grey–grey. Often individual particles varied in colour from centre to edge (*e.g.*, Fig. 29). In such cases the centre was invariably nearer to the yellow end of the series. These observations agree with those of Hume-Rothery, Raynor, and Little,[2] which were made on specimens cut from the bars used in the present work.* The nitride and duplex particles tended to be concentrated in clusters while the grey particles tended to be uniformly distributed.

(*b*) The Fe_2Ti phase, which occurred as rounded white particles (*see* Fig. 21).

(*c*) Small tan-coloured particles, often in the form of rods, which were thought to be titanium sulphide. This was confirmed by comparison with similar particles in specially prepared melts containing titanium and sulphur and by examination of their behaviour in polarized light, which has been described by Morrogh.[3]

(*d*) Dark-brown or black particles usually occurring near nitride clusters, possibly consisting of oxides (Figs. 22 and 23). Both particles occurred in globular form and the black also frequently occurred as a continuous string forming a " backbone " to a long cluster of titanium nitride–carbide particles.

The TiC and Fe_2Ti phases are, of course, " true " constituents, and the others are impurities.

Specimens were examined in the furnace-cooled condition, and after half-hour treatments *in vacuo* at temperatures in the range 700°–1200° C. followed by water-quenching. Vickers diamond-pyramid-hardness tests were also made on these specimens and the results, given in Table II, assisted in locating phase boundaries. The results of thermal analysis at heating and cooling rates of about 7° C./min. are given in Table III and of experiments with the Chevenard dilatometer at heating and cooling rates of about 20° C./min. in Table IV. All the results are in satisfactory agreement and vertical sections of the Fe–Ti–C diagram at different titanium contents, based on them, are given in Fig. 1 ((*a*)–(*h*)). In locating the position of the point *P* the results of quench-ageing tests, described later, were taken into account. The limits of existence of the titanium carbide phase could not be determined accurately owing to the uncertainty of composition engendered by the occurrence of the range of compounds from carbide to nitride, but approximate limits are shown by broken lines. Because of the short heating time, the sections probably do not represent complete equilibrium.

The main effect of titanium is to displace the boundaries at the low-carbon side of the regions containing γ or Fe_3C to higher carbon contents, and to raise A_3 temperatures. These results are consequences of the affinity of titanium for carbon and of the γ loop in the Fe–Ti binary diagram. In addition there is a tendency for the A_1 temperature to rise with increase in titanium. As drawn, it is raised 10° C. by 2·3% of titanium. There also seems to be a small tendency for the A_2 temperature to be raised by titanium, to the extent of ½–1° C. per 1% of titanium.

These diagrams are similar to those of Tofaute and Büttinghaus[4] for the pure iron–carbon–titanium equilibrium diagram. In the pure system, however, the titanium carbide solubility

* It was shown that titanium carbide and titanium nitride form a continuous series of solid solutions, the colour changing from yellow-orange–brown-grey with increase in carbon content.

TABLE II—*Vickers Diamond Hardness of Quenched Specimens*

Mark	C, %	Ti, %	Temperature, ° C., at which Heated for ½ hr. *in vacuo* before Water-Quenching									
			700	800	850	900	950	1000	1050	1100	1150	1200
A 0·0	0·11	Nil	206	239	...	306	...	350	...	341	...	331
A 0·1	0·11	0·08	...	195	211	192	203	...	237	...	...	...
A 0·25	0·09	0·24	196	185	192	194	281	305	311	311	330	330
A 0·5	0·08	0·50	103	102	104	104	178	200	247	267	290	302
A 1·0	0·11	1·40	127	138	138	137	142	145	141	141	141	126
A 2·0	0·10	2·20	143	...	147	...	152	...	149	...	152	147
A 4·0	0·09	4·02	203	...	...	...	...	192	...	...	...	189
A 6·0	0·08	6·08	230	...	...	...	...	226	...	...	...	222
B 0·0	0·21	Nil	185	393	...	436	...	465	...	449	...	418
B 0·1	0·25	0·10	175	442	408	475	458	466	462	434	451	481
B 0·25	0·23	0·32	178	271	352	425	429	423	447	442	451	402
B 0·5	0·23	0·64	178	220	241	390	...	375	370	350	350	370
B 1·0	0·25	1·05	191	191	173	175	176	263	259	331	336	322
B 2·0	0·27	2·25	128	130	131	129	...	164	215	264	...	269
B 4·0	0·24	4·96	182	...	...	...	...	182	...	...	...	171
B 6·0	0·22	6·12	248	...	...	...	...	215	...	...	...	211
C 0·0	0·38	Nil	204	663	...	660	...	633	...	610	...	601
C 0·1	0·39	0·08	186	603	604	601	585	591	577	571	628	626
C 0·25	0·38	0·24	175	500	560	...	588	578	582	555	560	555
C 0·5	0·37	0·68	193	341	360	493	484	480	473	479	483	450
C 1·0	0·35	1·00	182	198	244	313	396	400	...	414	396	410
C 2·0	0·41	2·30	160	...	156	...	177	234	310	...	329	325
C 4·0	0·41	4·25	187	...	...	...	...	171	...	...	...	176
C 6·0	0·37	5·36	203	...	...	...	...	204	...	...	...	198
D 0·0	0·52	Nil	204	784	...	766	...	778	...	...	...	...
D 0·1	0·56	0·06	202	758	765	786	782	783	787	770	790	...
D 0·25	0·54	0·27	171	675	702	694	716	734	703	720	700	700
D 0·5	0·50	0·72	195	554	629	620	595	645	622	584	610	591
D 1·0	0·53	1·10	193	388	529	518	...	564	...	493	...	500
D 2·0	0·52	2·30	210	301	...	...	420	424	420	...	420	391
D 4·0	0·50	4·60	181	192	...	...	...	191	...	...	...	186
D 6·0	0·60	5·83	181	...	...	...	...	185	...	...	...	164
E 0·0	0·81	Nil	232	955	...	934	...	878	...	872	...	946
E 0·1	0·76	0·10	246	866	826	868	849	809	821	849	847	840
E 0·25	0·74	0·26	200	850	852	880	870	856	835	843	858	850
E 0·5	0·72	0·60	210	843	836	814	850	838	875	850	817	875
E 1·0	0·74	0·75	206	839	811	826	794	836	789	818	818	776
E 2·0	0·77	2·70	209	511	583	...	587	620	618	575	...	584
E 4·0	0·67	4·34	165	163	...	...	...	162	...	...	...	167
E 6·0	0·83	5·48	185	...	...	...	...	196	...	...	...	183
F 0·0	1·10	Nil	243	893	...	908	...	...	...	...	...	931
F 0·1	1·16	0·085	239	935	917	958	...	943	903	861	870	855
F 0·25	1·07	0·26	221	910	880	877	857	874	871	865	880	...
F 0·5	1·06	0·62	245	868	930	880	904	919	907	859	880	850
F 1·0	1·00	1·10	240	...	877	...	884	...	869	...	905	913
F 2·0	1·05	2·20	243	907	861	...	845	...	840	...	882	869
F 4·0	1·03	4·70	218	240	...	503	...	530	...	470	...	514
F 6·0*X*	1·13	6·06	175	162	...	...	...	176	...	207	...	218

TABLE III—*Results of Thermal Analysis*

(Rate of heating and cooling approximately 7° C./min.)

Mark	C, %	Ti, %	Ac_1	Ac_2	Ac_3	Maximum Temperature	Ar_3	Ar_2	Ar_1*	Mean Ac_3–Ar_3
A 0·0	0·11	Nil	734	764	879	980	830	759	663	761
A 0·1	0·11	0·08	742	763	892	1005	847	760	676	761
A 0·25	0·09	0·24	...	762	913	1000	847	756	...	...
A 0·5	0·08	0·50	...	762	941	1007	885	757	...	759
A 1·0	0·11	1·40	...	767	...	1004	...	764	...	765
A 2·0	0·10	2·20	...	766	...	1003	...	764	...	765
A 4·0	0·09	4·02	...	765	...	1045	...	767	...	766
A 6·0	0·08	6·08	...	765	...	1052	...	768	...	766
A 6·0*X*	0·05	5·02				Not tested				
B 0·0	0·21	Nil	742	766	843	965	789	...	676	...
B 0·1	0·25	0·10	737	760	843	962	784	761	679	760
B 0·25	0·23	0·32	742	762	868	1011	806	758	676	760
B 0·5	0·23	0·64	740	760	...	928	...	755	688–675	757
B 1·0	0·25	1·05	...	759	...	927	...	757	...	758
B 1·0*X*	0·23	0·85				Not tested				
B 2·0	0·27	2·25	...	760	...	968	...	756	...	758
B 4·0	0·24	4·96	...	761	...	1052	...	760	...	760
B 6·0	0·22	6·12	...	763	...	1048	...	764	...	763
C 0·0	0·38	Nil	733	...	...	929	743	...	680	...
C 0·1	0·39	0·08				Not tested				
C 0·25	0·38	0·24	737	761	816	924	753	...	690–679	...
C 0·5	0·37	0·68	741	761	...	926	801	761	678	761
C 1·0	0·35	1·00	721	...	...	927	...	756	690–663	...
C 2·0	0·41	2·30	...	759	...	927	...	755	...	...
C 4·0	0·41	4·25	...	758	...	1060	...	756	...	757
C 6·0	0·37	5·36	...	761	...	1059	...	761	...	761
D 0·0	0·52	Nil	732	...	...	929	728	...	685	...
D 0·1	0·56	0·06	734	...	777	929	721	...	697	...
D 0·25	0·54	0·27	736	761	797	925	727	...	693–678	...
D 0·5	0·50	0·72				Not tested				
D 1·0	0·53	1·10	740	...	...	929	784–770	...	686	...
D 2·0	0·52	2·30	742	760	...	927	...	756	678–655	...
D 4·0	0·50	4·60	...	763	...	1042	...	760	...	761
D 6·0	0·60	5·83	...	763	...	1045	...	760	...	761
E 0·0	0·81	Nil	729	...	...	929	...	...	692	...
E 0·1	0·76	0·10	737	...	752	928	...	...	706	...
E 0·25	0·74	0·26	735	...	765	925	...	...	699	...
E 0·5	0·72	0·60	737	...	767	925	707	...	699	...
E 1·0	0·74	0·75	738	...	790	927	716	...	695	...
E 2·0	0·77	2·70	748	...	800	926	799	...	693	...
E 4·0	0·67	4·34	...	755	...	1040	...	754	...	754
E 6·0	0·83	5·48	...	759	...	1048	...	756	...	757
F 0·0	1·10	Nil	734	...	...	927	...	...	705	...
F 0·1	1·16	0·085	735	...	754	927	...	...	705	...
F 0·25	1·07	0·26	735	...	751	925	...	...	705	...
F 0·5	1·06	0·62	737	752	764	925	...	...	699–663	...
F 1·0	1·00	1·10	738	...	776	929	...	...	699	...
F 2·0	1·05	2·20	742	...	779	928	704	...	697	...
F 4·0	1·03	4·70	759	...	777	1057	...	754	680	...
F 6·0	1·09	5·65				Not tested				
F 6·0*X*	1·13	6·06	...	754	...	1055	...	751	...	752

* A small extra arrest was sometimes found just below the main Ar_1 arrest, and accounts for the double values in this column.

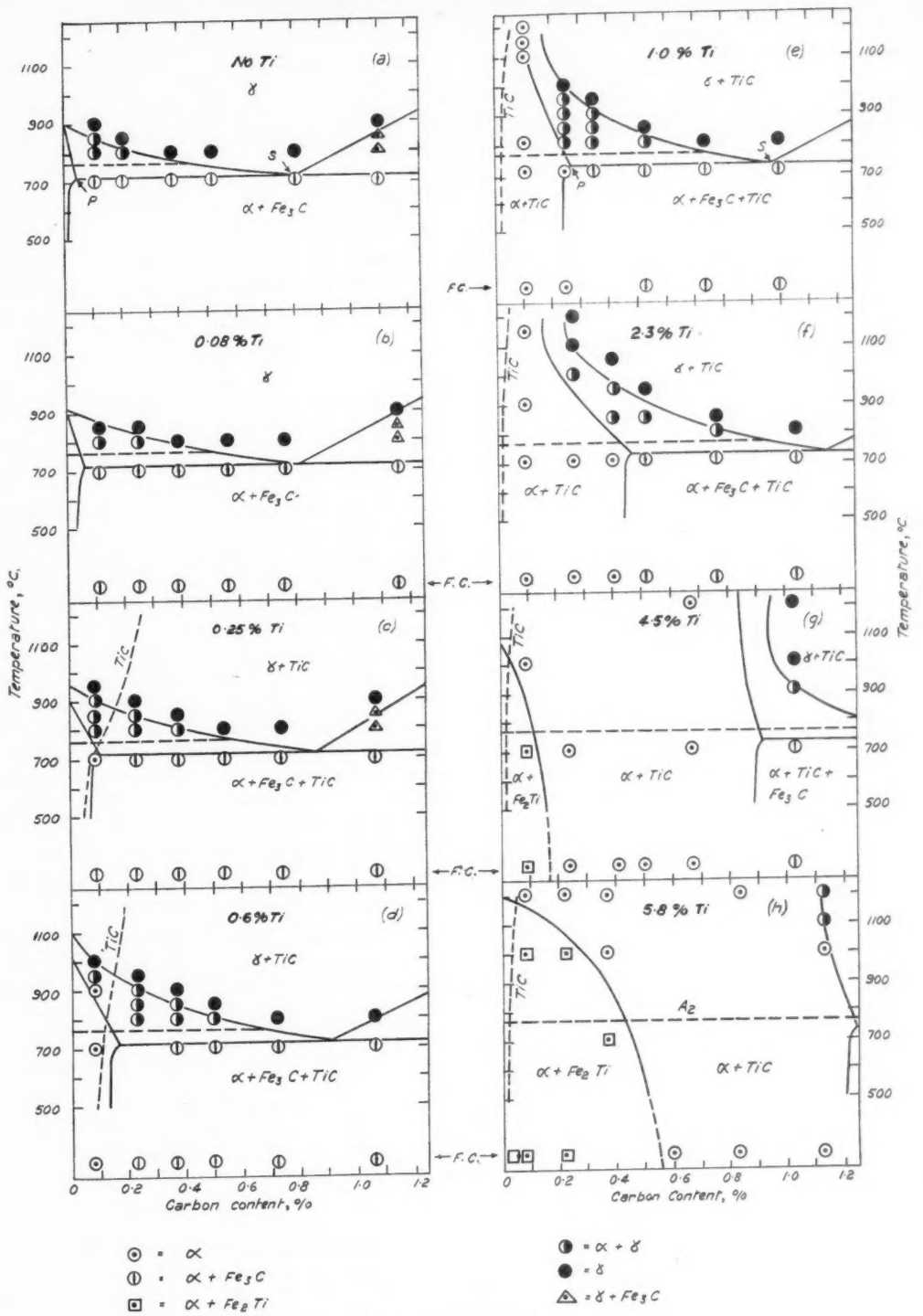


FIG. 1—Vertical sections of Fe–Ti–C constitutional diagram at various titanium contents. The broken lines indicate the TiC solid solubility and A_3 temperatures

TABLE IV—*Temperatures of Dilatometer Arrests*

Specimen	Composition		Ac_1, ° C.	Ac_3, ° C.	Maximum Temperature, ° C.	Ar_3, ° C.	Ar_1, ° C.
	C, %	Ti, %					
B 0·5	0·23	0·64	750	900	1000	840	670
C 1·0	0·35	1·00	750	905	990	850	680
D 0·0	0·52	Nil	728	780	970	705	670
D 2·0	0·52	2·30	755	913	1005	860	670
E 1·0	0·74	0·75	745	...	1005	...	690
E 4·0	0·67	4·34	...	...	...	...	...
F 2·0	1·05	2·20	...	765	1010	755	...
F 4·0	1·03	4·70	765	905	1015	860	680

limit occurs at considerably higher carbon contents, and points *P* and *S* are at slightly higher carbon contents, than the positions given in Fig. 1.

The titanium-bearing specimens quenched from high temperatures were found to have a smaller grain-size (presumably the austenite grain-size) than the corresponding titanium-free specimens. The effect was confirmed by standard McQuaid-Ehn tests, according to which the A.S.T.M. grain-size was reduced from 3 with no titanium to 6 with 0·5% titanium ; this effect may of course be due to aluminium.

Mechanical Properties

Tensile, Hardness, and Izod Tests in the Annealed Condition

Tensile test-specimens, with a parallel portion 1·4 in. long and 0·358 in. in dia., and standard Izod notched specimens were used for determining mechanical properties. The results are given in Table V. The large inclusion content, mainly of particles of titanium nitride type, of specimens containing 2–6% of titanium has considerably influenced the results in this titanium range. The fractured test-pieces showed defects, in extreme cases in the form of seams, and in the less serious cases in the form of powdery-looking areas. Photographs of particularly bad fractures are reproduced in Fig. 20 and the structure near the fracture of a tensile specimen containing 0·68% of titanium in Figs. 22 and 23.

Allowing for this factor, the effect of titanium can be interpreted as a consequence of the formation of TiC, at the expense of Fe_3C, and Fe_2Ti when sufficient titanium is present. Thus the maximum Izod value for each series occurs at the titanium/carbon ratio of approximately 4:1 (atomic ratio 1 : 1), corresponding to a microstructure of ferrite with very little titanium in solid solution, fairly uniformly distributed TiC particles, and no pearlite. With considerable titanium in excess of this ratio a network of Fe_2Ti reduces the impact strength to a very low value. Ductility tends to follow the same sequence but is affected much more by the inclusion content and therefore behaves less regularly. The ultimate tensile strength and hardness of the low-carbon series rise with increasing titanium content, and in the high-carbon series they fall with increasing titanium. In general no beneficial effect on mechanical properties of annealed specimens is brought about by titanium.

Influence of Titanium and Atmosphere on Hardness after Quenching from 700°–1200° C.

The results for specimens quenched from vacuum have already been given in Table II and the typical effect is illustrated in Fig. 2. Titanium

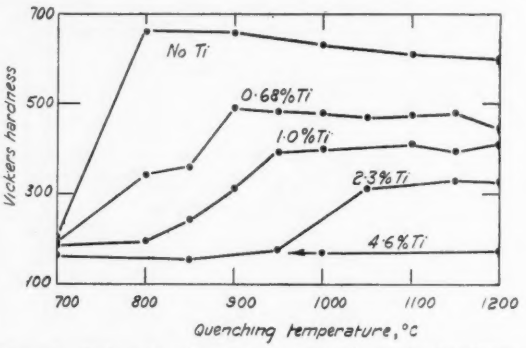


FIG. 2—Influence of titanium content on the hardness/quenching-temperature curves for steels containing 0·35% of carbon

TABLE V—*Mechanical Properties*

(Specimens annealed for ½ hr. at 900° C. and furnace-cooled)

Mark	Composition		Mechanical Properties					
	C, %	Ti, %	Ultimate Tensile Stress, tons/sq. in.	0·2% Proof Stress	Elongation on 1 in., %	Reduction of Area, %	Izod Impact Value	V.D.H.
A 0·0	0·11	Nil	25·0	17·5	51	69	81	107
A 0·1	0·11	0·08	23·35	18·0	54	75	89	108
A 0·25	0·09	0·24	25·75	17·0	52	75	92	106
A 0·5	0·08	0·50	29·7	24·0	26	75	84	117
A 1·0	0·11	1·40	28·32	11·9	46	62	92·5	126
A 2·0	0·10	2·20	30·2	13·3	38	51	42·5	158
A 4·0	0·09	4·02	42·0	...	22	45	2·9	195
A 6·0	0·08	6·08	43·9	...	18	23	2·1	218
B 0·0	0·21	Nil	29·35	18·5	44	62	51	123
B 0·1	0·25	0·10	31·45	...	40	60	62	140
B 0·25	0·23	0·32	28·3	15·2	40	70	80	145
B 0·5	0·23	0·64	29·2	18·7	44	69	93	136
B 1·0	0·25	1·05	28·0	27·2	38	68	100	137
B 1·0*X*	0·23	0·85	30·1	26·0	60	72	104	150
B 2·0	0·27	2·25	26·4	12·0	30	56	43·8	146
B 4·0	0·24	4·96	33·6	18·6	28	32	32·3	171
B 6·0	0·22	6·12	28·8	...	*	*	10·2	236
C 0·0	0·38	Nil	37·5	21·0	33	51	24	154
C 0·1	0·39	0·08	35·6	...	37	53	30	147
C 0·25	0·38	0·24	32·9	14·2	32	57	39	167
C 0·5	0·37	0·68	32·1	15·0	29	47	68	159
C 1·0	0·35	1·00	32·4	22·0	36	56	67	151
C 2·0	0·41	2·30	27·0	19·0	33	65	53	175
C 4·0	0·41	4·25	31·8	18·9	18	21	45	174
C 6·0	0·37	5·36	21·2	...	*	*	7	204
D 0·0	0·52	Nil	42·4	25·0	29	43	13·9	173
D 0·1	0·56	0·06	42·25	...	29	40	12·9	179
D 0·25	0·54	0·27	37·0	14·4	33	51	25·5	180
D 0·5	0·50	0·72	36·7	17·0	33	53	33·1	177
D 1·0	0·53	1·10	34·7	17·0	33	51	43·2	184
D 2·0	0·52	2·30	33·9	16·2	32	47	47·3	180
D 4·0	0·50	4·60	34·3	14·0	10	17	32·8	170
D 6·0	0·60	5·83	29·3	...	*	*	37·8	181
E 0·0	0·81	Nil	50·9	25	16	20	6·9	232
E 0·1	0·76	0·10	50·8	22	22	29	4·7	229
E 0·25	0·74	0·26	44·6	16·0	23	36	8·5	223
E 0·5	0·72	0·60	44·9	16·5	22	35	14·8	226
E 1·0	0·74	0·75	44·9	16·5	26	36	9·5	224
E 2·0	0·77	2·70	38·9	16·7	22	27	33·4	204
E 4·0	0·67	4·34	30·5	17·1	17	23	23·9	151
E 6·0	0·83	5·48	32·2	...	10	12	26·1	178
F 0·0	1·10	Nil	57·2	>35	12	14	3·8	237
F 0·1	1·16	0·085	56·0	33	13	12	5·5	245
F 0·25	1·07	0·26	49·0	18·5	16	20	5·9	248
F 0·5	1·06	0·62	53·6	17·0	12	13	6·6	266
F 1·0	1·00	1·10	47·8	17·0	17	20	3·0	266
F 2·0	1·05	2·20	53·7	23·0	10	9	5·3	280
F 4·0	1·03	4·70	34·2	26·0	*	*	44·6	196
F 6·0*X*	1·13	6·06	39·0	...	14	15	19·8	170

* Could not be fitted together.

TABLE VI—*Hardness of Quenched Specimens Heated in Air or Vacuum*

Specimen	Composition		Quenching Temperature, ° C.	Method of Heating in Air	V.D.H. Air-Heated	V.D.H.* Vacuum-Heated	Percentage Change in Hardness on Air-Heating
	C, %	Ti, %					
C 1·0	0·35	1·00	1000	Electric muffle	220	400	−45
C 1·0	0·35	1·00	1000	Gas furnace	420	400	+ 5
C 1·0	0·35	1·00	1000	Electric tube furnace	414	400	+ 3
C 0·5	0·37	0·68	900	Electric muffle	444	493	−10
E 0·0	0·81	Nil	900	,, ,,	833	934	−11
E 1·0	0·74	0·75	900	,, ,,	801	826	− 3
E 2·0	0·77	2·70	900	,, ,,	582	585	− 1
C 0·5	0·37	0·68	1000	Electric muffle	454	480	− 5
E 0·0	0·81	Nil	1000	,, ,,	798	878	− 9
E 1·0	0·74	0·75	1000	,, ,,	798	800	0
E 2·0	0·77	2·70	1000	,, ,,	452	620	−27
C 0·5	0·37	0·68	1100	Electric muffle	502	479	+ 5
E 0·0	0·81	Nil	1100	,, ,,	834	872	− 4
E 1·0	0·74	0·75	1100	,, ,,	469	800	−41
E 2·0	0·77	2·70	1100	,, ,,	320	580	−45

* Where actual figure not available in Table II the value here was obtained by interpolation.

lowers the maximum hardness on account of the reduced carbon content of the γ phase, and raises the quenching temperature required to attain it.

Variable results were obtained when specimens were heated in air instead of *in vacuo* before quenching, so trials were carried out with a selection of the steels to compare the effect of different methods of heating ; the results are summarized in Table VI. The preliminary experiments with the 0·35% carbon, 1·0% titanium steel, *C* 1·0, in the different furnaces showed that a considerable drop in hardness was experienced when the heating was carried out in an electric muffle, where the oxidizing conditions might be expected to be more pronounced than in a gas furnace or closed tube furnace. Subsequently experiments on the effect were therefore made using only the electric muffle for heating in air. The differences were thought to be due to the formation on the titanium steels when heated in air of an oxide film having a greater insulating effect during quenching than the normal oxide layer formed on the straight carbon steels. From the results given in Table VI it would appear that a short treatment at 900° C. is insufficient to form much of a protective layer on either the titanium steels or the plain carbon steels. At higher temperatures the position is altered in that the titanium steels have a much lower hardness when heated in air than when heated *in vacuo*, but there is little difference in hardness between the differently treated plain carbon steels. The effect is most pronounced at the highest temperature used, namely, 1100° C.

Effect of Tempering

Many of the alloys were quenched from 1235° or 1000° C. after heating for 1 hr. *in vacuo* at these temperatures and then tempered at increasing temperatures from 200° to 800° C. to determine the influence of tempering upon hardness ; the results are given in Table VII.

It is possible to distinguish two separate main effects of titanium. Steels with a low titanium/carbon ratio soften when tempered in accordance with the normal decomposition of martensite when it is heated even to moderate temperatures, but the presence of 1% or more of titanium tends to stabilize the martensite so that its decomposition is delayed at temperatures as high as 600° C. This effect is illustrated in Figs. 3–5. With a high titianium/carbon ratio and appreciable amounts of titanium, the hardness of the quenched steel may be increased on tempering owing to precipitation-hardening by Fe_2Ti. TiC can also cause precipitation-hardening.

Retention of the martensitic hardness is decidedly more marked after quenching from 1235° C. than after quenching from 1000° C. The optimum Ti/C atomic ratio is a little greater than 1 : 1 (4 : 1 by weight), at least partly because some titanium is bound as nitride. With less titanium than the optimum, the hardness, although higher immediately after quenching, drops more

rapidly on tempering, and with more titanium the initial hardness is lower so that in both cases the hardness after tempering at 600° C. is less than with the optimum ratio. With the 1 : 1 atomic ratio, the as-quenched hardness is almost completely maintained on tempering for one hour

TABLE VII—*Influence of Titanium on the Effect on Hardness of Tempering Quenched Steels*

Specimen	Composition		V.D.H., after Quenching	V.D.H., after Tempering for 1 hr. at:				Percentage Change in V.D.H. after Tempering at 600° C.
	C, %	Ti, %		200° C.	400° C.	600° C.	800° C.	
			Softening of Martensite after Quenching from 1000° C.					
B 0·0	0·21	Nil	364	354	291	207	120	−43
B 0·1	0·25	0·10	383	378	303	247	127	−36
B 0·25	0·23	0·32	415	379	307	264	116	−36
B 1·0*X*	0·23	0·85	231	215	205	208	...	−10
B 2·0	0·27	2·25	197	192	181	198	...	+ 5
C 0·5	0·37	0·68	447	409	326	271	254	−39
C 1·0	0·35	1·00	390	397	315	284	145	−27
C 2·0	0·41	2·30	241	274	254	245	...	+ 2
D 2·0	0·52	2·30	424	438	...	297	153	−30
			Softening of Martensite after Quenching from 1235° C.					
B 0·0	0·21	Nil	394	377	308	210	...	−47
B 0·1	0·25	0·10	449	436	367	291	...	−35
B 0·25	0·23	0·32	414	425	379	350	...	−15
B 1·0*X*	0·23	0·85	361	363	345	359	203	− 1
B 2·0	0·27	2·25	260	356	268	252	...	− 3
C 0·0	0·38	Nil	606	544	360	252	...	−58
C 0·25	0·38	0·24	608	524	362	320	...	−47
C 0·5	0·37	0·68	478	464	376	347	...	−27
C 1·0	0·35	1·00	430	442	380	370	...	−14
C 2·0	0·41	2·30	338	341	319	330	...	− 2
D 0·5	0·50	0·72	634	590	380	310	...	−51
D 2·0	0·52	2·30	442	448	388	372	...	−16
D 4·0	0·50	4·60	177	175	180	188	182	+ 6
D 6·0	0·60	5·83	164	172	171	182	172	+11
F 0·0	1·10	Nil	862	736	468	318	...	−63
F 0·25	1·07	0·26	888	752	480	364	...	−59
F 0·5	1·06	0·62	898	766	496	372	...	−59
F 2·0	1·05	2·20	886	748	492	370	...	−58

Precipitation-Hardening by Fe_2Ti, Quenched from 1235° C.

				V.D.H. after Tempering for 1 hr. at:					Percentage Change in V.D.H. after Tempering at 700° C.
				200° C.	400° C.	600° C.	700° C.	800° C.	
A 1·0	0·11	1·40	142	146	138	148	140	134	− 1
A 2·0	0·10	2·20	161	159	160	164	165	149	+ 2
A 4·0	0·09	4·02	181	189	183	201	203	201	+12
A 6·0	0·08	6·08	222	224	227	245	320	278	+44

TABLE VIII—*Effect of Titanium on Quench-Ageing*

Specimen	Composition		Annealed 900° C. ½ hr., Furnace-Cooled, V.D.H.	V.D.H.* after Ageing at Room Temperature for :												
	C, %	Ti, %		1 hr.	1 day	Increase over 1 hr.	1 week	Increase over 1 hr.	1 month	Increase over 1 hr.	2 months	Increase over 1 hr.	10 months	Increase over 1 hr.	68 months	Increase over 1 hr.
Quenched from 685° C. after ½ hr. at 685° C.																
A 0·0	0·11	Nil	114	151	164	13	198	47	203	52	205	54	199	48	187	36
A 0·1	0·11	0·08	112	147	159	12	194	47	199	52	201	54	185	39	184	38
A 0·25	0·09	0·24	136	169	171	2	181	12	194	25	196	27	185	16	175	6
A 0·5	0·08	0·50	104	110	107	−3	107	−3	110	0	108	−2	110	0	106	−4
A 1·0	0·11	1·40	131	147	141	−6	148	1	144	−3	136	−11	149	2	134	−13
A 2·0	0·10	2·20	150	166	162	−4	157	−9	151	−15	153	−13	153	−13	150	−16
B 0·25	0·23	0·32	132	166	171	4	171	5	193	27	190	24	183	17	175	9
B 0·5	0·23	0·64	128	186	188	2	196	10	212	26	208	22	188	2	187	1
B 1·0*X*	0·23	0·85	154	178	180	2	186	8	196	18	190	22	194	17	180	3
B 2·0	0·27	2·25	137	131	133	2	118	−13	129	−2	122	−9	125	−6	126	−4
C 1·0	0·35	1·00	157	191	191	0	200	9	194	3	217	26	209	18	196	5
C 2·0	0·41	2·30	173	169	170	1	174	5	167	−2	181	12	200	31	172	3
D 2·0	0·52	2·30	166	184	193	9	191	7	189	5	195	11	205	21	195	11
E 2·0	0·77	2·70	181	212	215	3	213	1	226	14	253	41	244	32	208	−4
F 2·0	1·05	2·20	262	254	255	1	256	2	254	0	233	−21	270	16	240	−14
Air-cooled from 900° C., ½ hr. at 685° C., and quenched																
A 0·0	0·11	Nil	125	154	163	9	198	44	194	40	199	45	190	36	183	29
Air-cooled from 900° C., 4 hr. at 685° C., and quenched																
A 0·0	0·11	Nil	124	141	151	10	190	49	190	49	197	56	187	46	179	38

* Each hardness figure represents the average of three or more impressions.

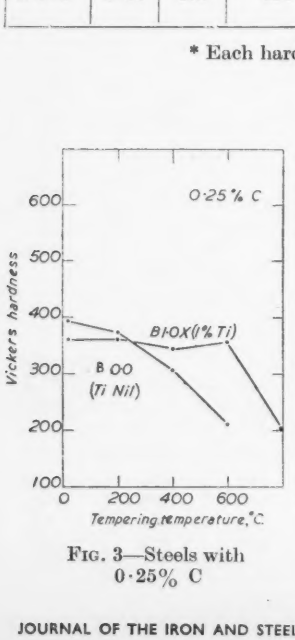


FIG. 3—Steels with 0·25% C

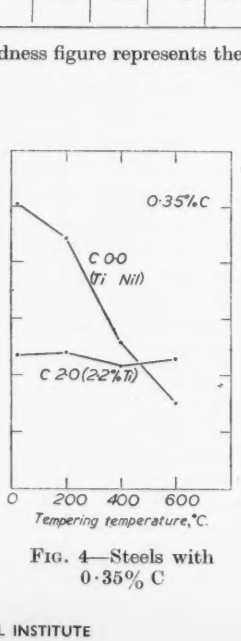


FIG. 4—Steels with 0·35% C

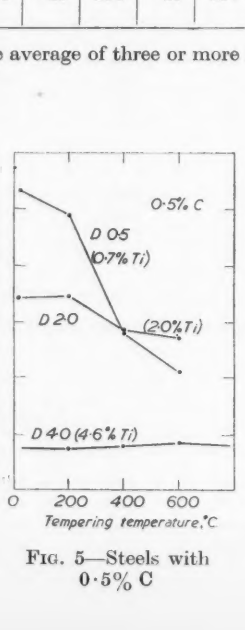


FIG. 5—Steels with 0·5% C

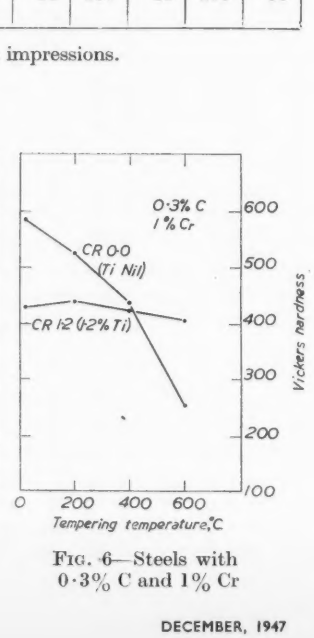


FIG. 6—Steels with 0·3% C and 1% Cr

at 600° C. if the quenching temperature was 1235° C., as steels *B* 1·0*X* and *C* 2·0 show.

It seems that greater retention of hardness after quenching from 1235° C. must be a consequence of increased solution of TiC, and can hardly be entirely due to precipitation-hardening by TiC being superimposed on the normal softening process, as has been suggested.[5] Two of the compositions in Table VIII, *D* 4·0 and *D* 6·0, which contain considerable quantities of TiC together with sufficient titanium to be in the α-phase region at all temperatures up to 1235° C., would show any precipitation-hardening due to TiC unobscured by other changes. However, in these steels the hardness increment on tempering at 600° C. is less than 20 units, whereas a minimum increment of 150 units would be required to offset the normal drop on tempering martensite at 600° C.

" Retentivity " seems more likely to be due to delayed drecipitation of carbide, as has also been suggested.[6] Confirmation was provided by microexamination of specimens *C* 0·0 and *C* 2·0 which possessed a normal acicular martensitic structure after quenching. The microstructures after tempering at 600° C. are shown in Figs. 26 and 29. Steel *C* 0·0 has a normal tempered structure with fairly uniformly dispersed precipitate, but *C* 2·0 shows no general precipitate even on deep etching as in the figure. This indicates that the carbon is still in solid solution in the latter specimen owing, presumably, to the considerable mutual affinity of titanium and carbon. A carbon atom diffusing through the iron lattice will tend to lock in place when it reaches a position adjoining a titanium atom to form a TiC molecule and *vice versa*. Subsequent diffusion of the titanium or carbon atom involves breaking the strong titanium–carbon bond or diffusion of the TiC molecule ; the probability of either process will be small at temperatures involved in tempering. The formation and precipitation of carbide aggregates are thus hindered and the normal softening of martensite on tempering consequently prevented.

The X-ray photographs, Figs. 27, 28, 30, and 31 of *C* 0·0 and *C* 2·0 agree with this hypothesis. In the quenched state the diffraction lines (Figs. 27 and 30), are diffuse, especially those of *C* 0·0 owing to the larger amount of carbon in solution. (*C* 2·0 gave also lines due to TiC and TiN). After tempering at 600° C. the diffraction lines of *C* 0·0 (Fig. 28) are sharp and a set due to cementite has appeared, but the lines of *C* 2·0 (Fig. 31) are nearly as diffuse as before tempering and no cementite lines can be detected. Little relief of lattice strain has therefore occurred in the latter specimen, which confirms that carbon is still in solution.

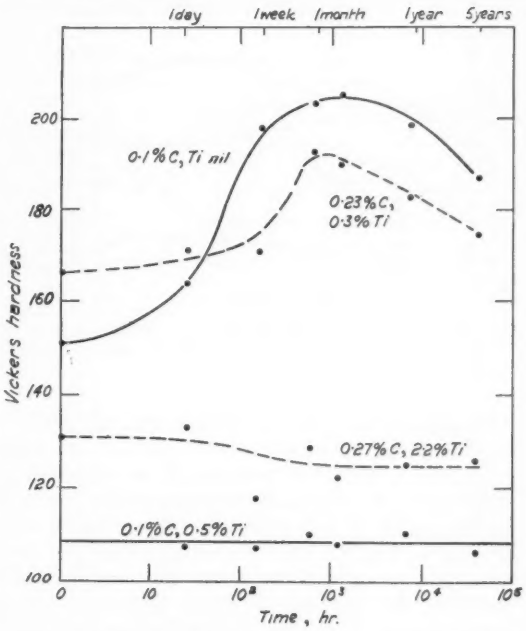


FIG. 7—Influence of titanium content on quench-ageing of steel water-quenched from 685° C.

Precipitation-hardening by Fe_2Ti is to be expected from the shape of the solubility curve of Fe_2Ti in iron. The solubility has been found to range from about 2% of titanium at atmospheric temperatures to 7% at the eutectic temperature of 1300° C.*[7] so that more than 2% of titanium must be present in the purer binary alloys before precipitation-hardening can occur. With the less pure steels it was thought some hardening might occur with less than 2% of titanium. The results in Table VII show, however, no definite hardening with 2·2% of titanium and only slight hardening with 4%, but there is an appreciable increase with 6%.

* This solubility curve was confirmed during the present investigation.

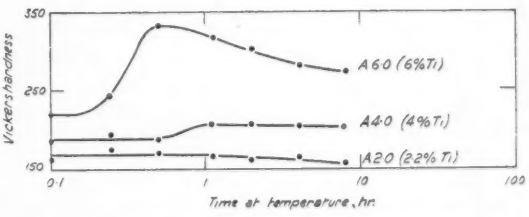


FIG. 8—Precipitation-hardening in steels containing 0·1% C, solution-treated at 1235° C. and tempered at 700° C.

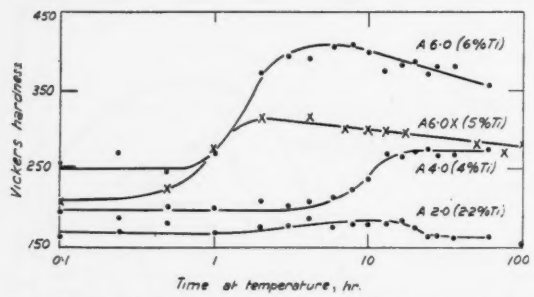


FIG. 9—Precipitation-hardening in steels containing 0·1% C, solution-treated at 1235° C. and tempered at 650° C.

The effect of tempering between 600° and 700° C. for increasing time upon the hardness of four compositions is shown graphically in Figs. 8–10. Steels *A* 2·0, *A* 4·0, and *A* 6·0 were solution-treated for 1 hr. at 1235° C. and water-quenched, and *A* 6·0*X* for 4 hr. at 1150° C. and water-quenched. Micro-examination showed that in the as-quenched condition no undissolved Fe_2Ti was present in any specimen. The curves show the usual characteristics of precipitation-hardening. There is slight hardening with 2·2% of titanium, and the maximum increase occurs with the 6% steel, *A* 6·0, treated at 650° C., when the hardness is raised from 250 to 405 in 5 hr. Figure 9 shows that *A* 6·0*X* hardened faster than *A* 4·0 or *A* 6·0, suggesting that incomplete homogenization of dissolved Fe_2Ti occurs in the solution treatment at 1150° C.

Quench-Ageing

It is known that titanium tends to eliminate the hardening and embrittling action occurring on ageing low-carbon steels rapidly cooled from 650°–700° C. It is assumed that the carbon mostly occurs as titanium carbide, which is insoluble in iron except at very high temperatures and is therefore normally out of solution and cannot be precipitated during ageing. The results of tests carried out with a number of the present steels are given in Table VIII and typical results are plotted in Fig. 7.

The effect of titanium in eliminating the hardness increase due to quench-ageing in low-carbon steels is confirmed ; the titanium also reduces the rate at which quench-ageing occurs. Jones[8] concluded that the titanium content should be about four times the carbon content, the ratio found in TiC, but the present experiments indicate that in order to eliminate completely increases in hardness due to ageing, there should be an excess of titanium over this, say six times the carbon content. It is possible that part of this titanium in excess of the TiC ratio may be required for combination with the nitrogen present.

Izod impact tests were performed on some of the compositions of Table VIII after furnace-cooling and after a half-hour anneal at 685° C., water-quenching, and ageing at room temperature. The results, each the average of three tests, are given in Table IX.

The results confirm previous work[8] in that a sufficient titanium content, approaching six times the carbon content, prevents quench-ageing embrittlement.

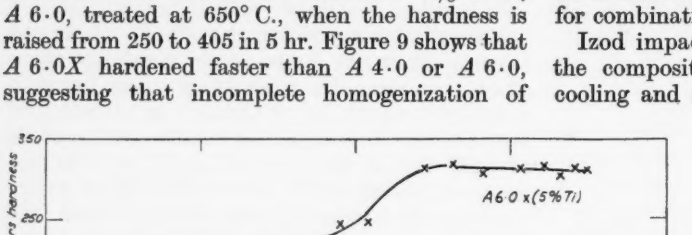


FIG. 10—Precipitation-hardening in steels containing 0·1% C, solution-treated at 1235° C., and tempered at 600° C.

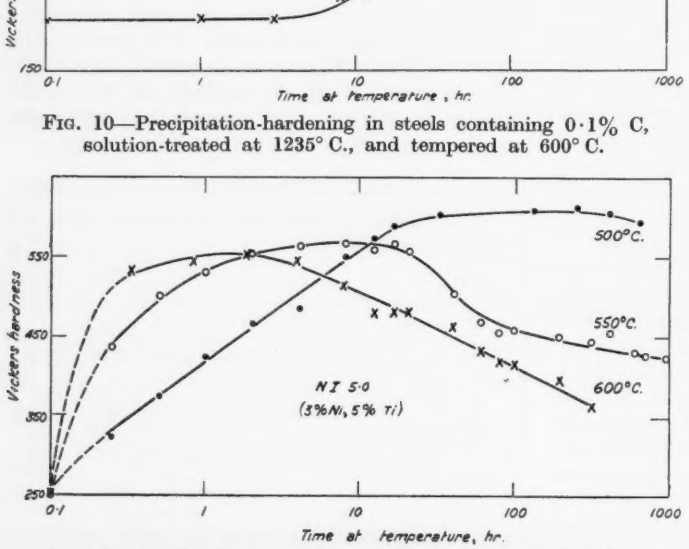


FIG. 11—Precipitation-hardening in steel *NI* 5·0, solution-treated at 1000° C. and then tempered at 500°, 550°, and 600° C.

Strain-Ageing

Five steels were tested for strain-ageing. Tensile specimens were elongated 10%, aged the same day for 1 hr. at 250° C. (other workers have shown this to give about the maximum effect[9]), and tested to destruction the following day. Results were calculated on the unstrained dimensions and are given in Table X, together with the comparative results on unaged material from Table V.

It appears that the first three steels have strain-aged and the last two have not, although the drop in ultimate tensile strength for *A* 0·5 is anomalous. As in the

TABLE IX—*Effect of Titanium on Quench-Ageing Embrittlement*

Steel	Composition		Izod Notched-Bar Impact Value		
	C, %	Ti, %	As Furnace-Cooled from 900° C.	Annealed at 685° C., Water-Quenched, and Aged at Room Temperature	
				For 1 day	For 1 year
A 0·0	0·11	Nil	81 u*	75 u	10
A 0·1	0·11	0·08	89 u	96 u	28
A 0·25	0·09	0·24	92 u	110 u	23
A 0·5	0·08	0·50	84	95 u	97 u
B 0·25	0·23	0·32	80 u	66 u	34
B 0·5	0·23	0·64	91	93 u	39

* u= Unbroken.

case of the previous ageing phenomena, with 0·1% of carbon a titanium content nearly six times as great is required to prevent strain-ageing.

Tensile Tests at Elevated Temperatures

The tensile tests at elevated temperatures were performed on temper-hardening (*A* 6·0*X*), "retentive" (*B* 1·0*X*, *C* 2·0, and *D* 2·0), and, for comparison, plain carbon steels (*A* 0·0, *C* 0·0, and *F* 0·0). The preliminary heat-treatment was varied to suit the composition; the temper-hardening specimens were precipitation-treated to produce approximately the maximum hardness and the "retentive" and plain carbon compositions were heated for 1 hr. at 600° C. to stabilize their structures. The retentive steels were given a preliminary solution treatment at and quenched from 1235° C. to develop the maximum retention of hardness, but the comparative plain carbon steel, *F* 0·0, was quenched from 1000° C. to reduce the risk of cracking. Steel *A* 0·0 was also quenched from this temperature. As these compositions contain no titanium the different quenching temperature should not have influenced the results appreciably. Specimens of 0·358-in. dia. were employed. They were raised to testing temperature in the tensile machine, held constant for 15–20 min. to ensure heat equilibrium, and strained at the rate of approximately $\frac{1}{15}$ in./min. The ultimate tensile stresses are plotted in Fig. 12 with a logarithmic ordinate in order to facilitate a satisfactory comparison of the slopes of the curves. Elongation and reduction of area showed in general the inverse behaviour of the ultimate tensile stress, except for some specimens containing seams similar to those in Fig. 20.

Of the non-precipitation-hardening steels those containing titanium not only possess a higher strength than the plain carbon steels, corresponding to their higher hardness, but show a more gradual decrease of strength with increase of temperature than do the latter. This latter property is more evident with steels *B* 1·0*X* and *C* 2·0 having "effective" titanium/carbon atomic ratios of 1 : 1 or greater, than with steel *D* 2·0 which has previously been suggested to have an effective ratio of less than 1 : 1.

The precipitation-hardening steel *A* 6·0*X* shows the least drop in strength with rise in temperature. Micro-examination and hardness testing on the ends of the tensile specimens showed, however, that the preliminary precipitation treatment had for some reason been inadequate to develop full hardness, so some and possibly all of the superiority of the *A* 6·0*X* composition in this respect is due to the precipitation-hardening process continuing during the tests, to a greater degree the higher

TABLE X—*Effect of Titanium on Strain-Ageing*

Steel	Composition		Stress for 10% Elongation, tons/sq. in.			Ultimate Tensile Strength, tons/sq. in.		
	C, %	Ti, %	Unaged	Aged	Increase	Unaged	Aged	Increase
A 0·0	0·11	Nil	23·0	27·2	4·2	25·0	29·3	4·3
A 0·1	0·11	0·08	23·0	25·4	2·4	23·3	26·3	3·0
A 0·25	0·09	0·24	26·3	29·5	3·3	25·7	30·1	4·4
A 0·5	0·08	0·50	21·7	21·5	−0·2	29·7	23·2	−6·5
A 1·0	0·11	1·40	27·3	27·5	0·2	28·3	28·9	0·6

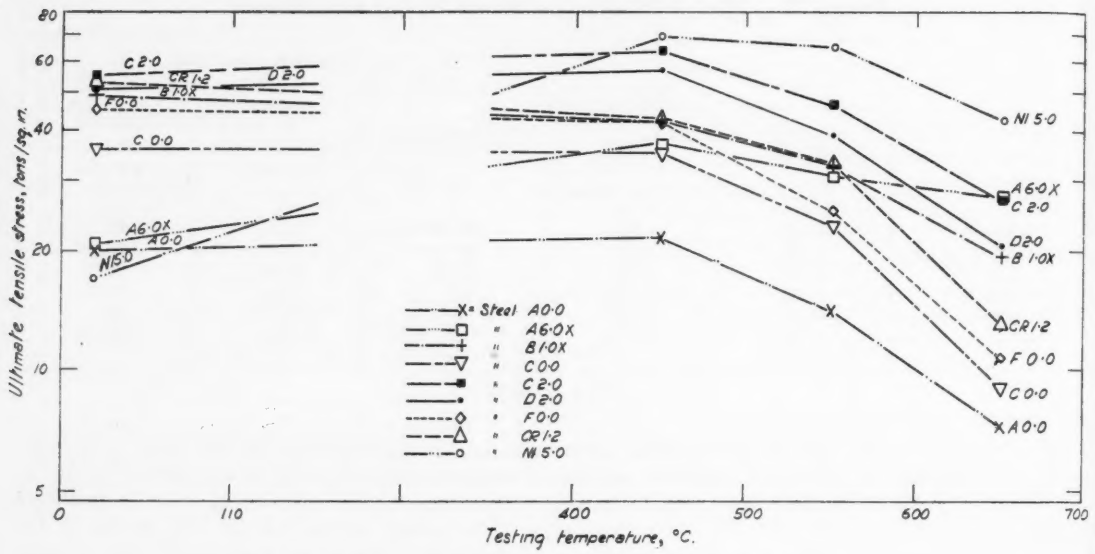


FIG. 12—Influence of temperature in short-time tensile tests

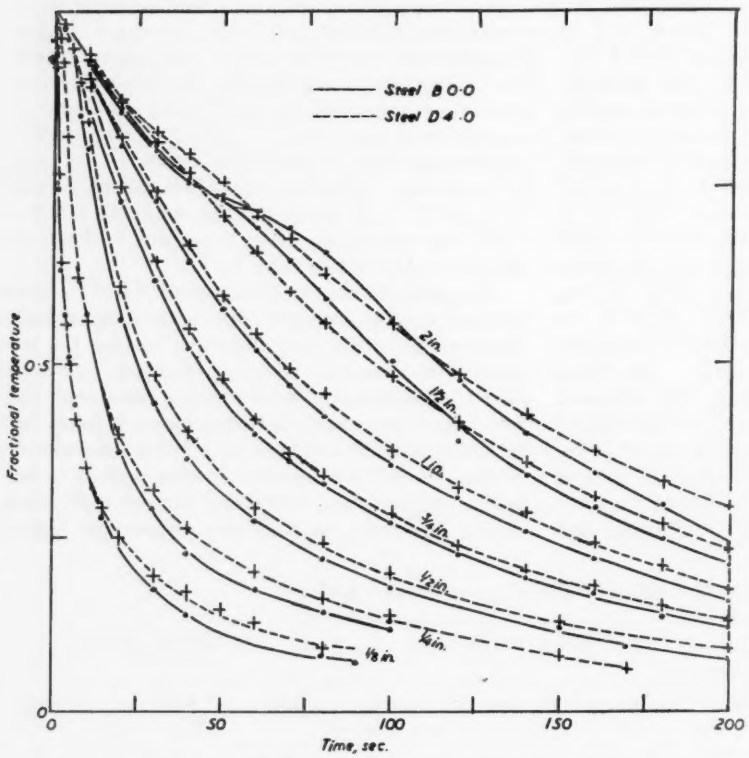


FIG. 13—Cooling curves at indicated distances from quenched ends of Jominy bars of steels *B* 0·0 (0·21% C, Ti nil) and *D* 4·0 (0·50% C, 4·6% Ti). Fractional temperature = (Instantaneous temperature — quenching-water temperature)/(Temperature at start of quench — quenching-water temperature)

the testing temperature. The room-temperature strength of this steel is anomalously low.

Hardenability

Jominy end-quench hardenability tests were performed on the *B* and *D* series (0·25 and 0·5% of carbon respectively), omitting the 6% titanium compositions. Since heating temperature was clearly an important factor, specimens of each composition were heated for 1 hr. at and quenched from two temperatures, 1000° C. and 1235° C. Specimens of the retentive steel *D* 2·0, of *D* 0·5, and of the comparative titanium-free steel *D* 0·0 were also treated for 1 hr. at 900° C. and 1100° C.

Two departures from standard Jominy practice were made : The bars were ½ in. in dia. and 3 in. long—*i.e.*, smaller than the standard 1-in. × 4-in. bar—since all specimens had to be obtained from a limited length of ⅝-in. dia. bar ; and after treatment all bars were slit axially

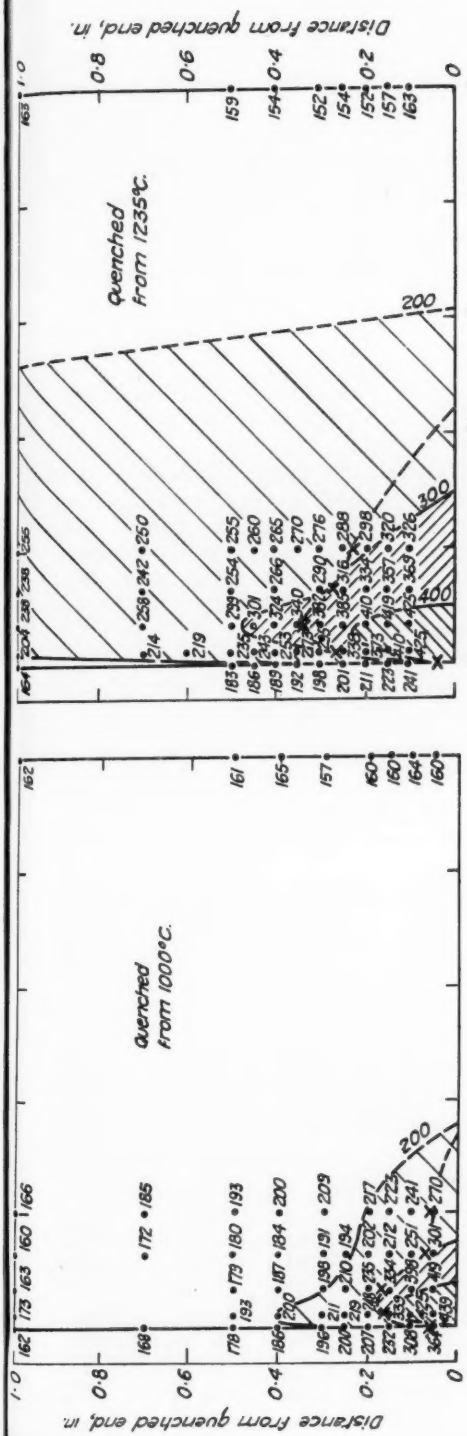


FIG. 14—Series *B*, 0·25% C, quenched from 1000° C.

FIG. 15—Series *B*, 0·25% C, quenched from 1235° C.

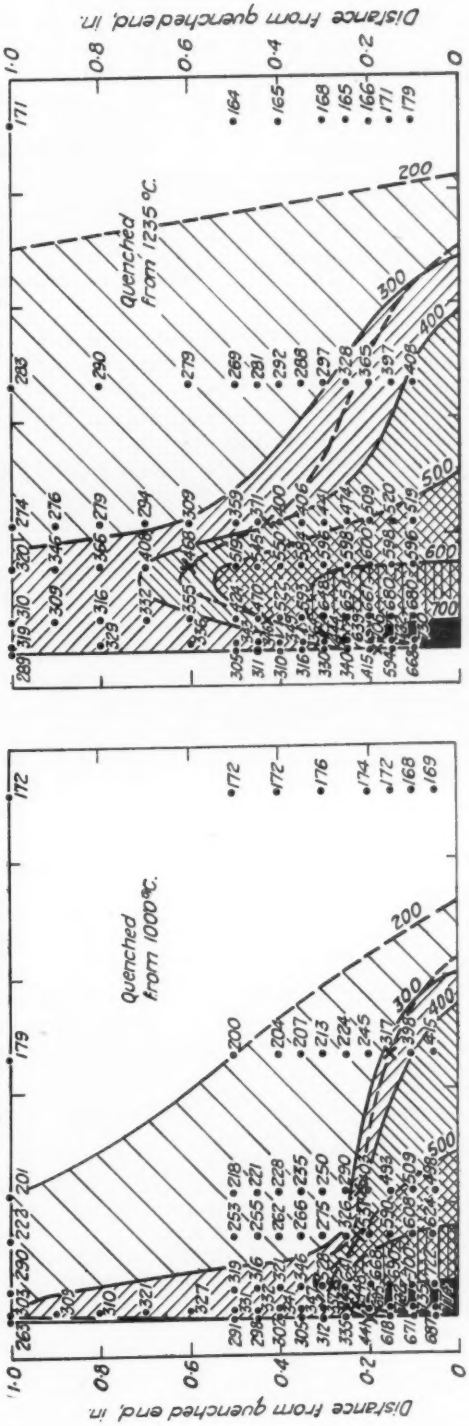


FIG. 16—Series *D*, 0·5% C, quenched from 1000° C.

FIG. 17—Series *D*, 0·5% C, quenched from 1235° C.

FIGS. 14–17—Influence of titanium on hardenability

and duplicate hardness tests made $\frac{1}{8}$ in. from each edge instead of on standard flats 0·02 in. deep, since, despite vacuum-heating of all specimens, those treated at high temperatures suffered sufficient decarburization to render such hardness results valueless. On some specimens the hardness 0·05 in. from the quenched end was lowered by decarburization. To facilitate comparison between the present work and any subsequent work on standard bars, cooling curves taken for the two extreme compositions (*B* 0·0 and *D* 4·0) at various distances from the quenched end are given in Fig. 13.

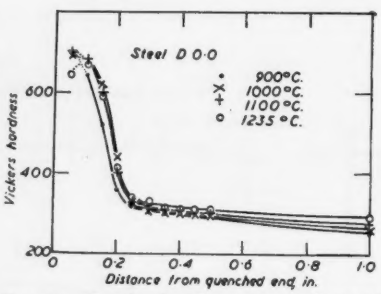


FIG. 18—Jominy hardness curves for steel *D* 0·0 (0·52% C, Ti nil), quenched from temperatures indicated

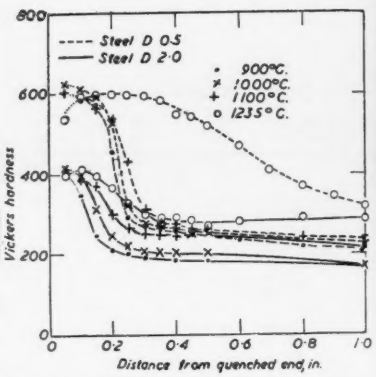


FIG. 19—Jominy hardness curves for steels *D* 0·5 (0·50% C, 0·72% Ti) and *D* 2·0 (0·52% C, 2·30% Ti). Bars quenched from temperatures indicated

The duplicate hardness results usually agreed to within ± 5 units and rarely deviated from a mean curve by more than ± 15 units. The averaged results are shown graphically in Figs. 14–19. The complete hardness curves for compositions *D* 0·0, *D* 0·5, and *D* 2·0 treated at different temperatures are given in Figs. 18 and 19, but to save space the results for the complete *B* and *D* series are given in the form of contour graphs in Figs. 14–17, with the distance from the quenched end to the point of inflection on the hardness curve, which is sometimes taken as a single-valued measure of hardenability, shown by a broken line. This line is drawn to cut the titanium axis at the point where, from Fig. 1, the γ phase just disappears at the temperature of treatment.

Figure 19 shows the expected increase in hardenability of the titanium steels *D* 0·5 and *D* 2·0 with increase in heating and quenching

TABLE XI—*Chemical Composition of Alloy Steels*

Mark	C, %	Ti, %	Cr, %	Mo, %	Mn, %	Ni, %	Si, %	Al, %
			Chromium–Molybdenum Steels					
CM 0·0	0·33	Nil	0·99	0·32	0·66	<0·005	0·15	<0·01
CM 0·25	0·39	0·27	1·09	0·30	0·61	...	0·18	...
CM 1·2	0·33	1·29	1·00	0·31	0·65	<0·005	0·25	0·29
CM 5·0	0·32	5·23	1·06	0·33	0·73	0·018	0·26	1·25
			Manganese–Molybdenum Steels					
MM 0·0	0·34	Nil	0·025	0·32	1·86	0·005	0·21	<0·01
MM 0·25	0·35	0·28	0·043	0·33	1·76	0·076	0·23	0·11
MM 1·2	0·38	0·86	0·15	0·31	1·86	0·005	0·08	0·19
MM 5·0	0·40	4·56	0·06	0·32	1·84	0·005	0·23	1·00
			Chromium Steels					
CR 0·0	0·32	Nil	1·13	Nil	0·61	0·02	0·22	<0·01
CR 0·25	0·31	0·32	1·06	Nil	0·69	0·02	0·22	0·08
CR 1·2	0·32	1·18	0·96	Nil	0·66	0·01	0·29	0·29
CR 5·0	0·40	4·97	1·06	Nil	0·78	0·03	0·12	1·00
			Nickel Steels					
NI 0·0	0·31	Nil	0·15	Nil	0·57	2·92	0·21	<0·01
NI 0·25	0·37	0·29	0·087	Nil	0·68	3·08	0·19	0·09
NI 1·2	0·39	1·02	0·06	Nil	0·73	2·86	0·28	0·21
NI 5·0	0·40	4·86	0·02	Nil	0·74	2·90	0·12	1·49

Fig. 20—Bad seams in tensile and Izod specimens of steel *B* 2·0 (0·27% C, 2·25% Ti)

Fig. 22 × 50

Fig. 23 × 500

Figs. 22 and 23—Steel *C* 0·5 (0·37% C, 0·68% Ti), near fracture of tensile specimen. Nitride clusters associated with dark-brown non-metallic inclusions

Fig. 24—No titanium, coarse structure × 2000

Fig. 25—1·2% Titanium, finer structure than Fig. 24. × 2000

Figs. 24 and 25—Chromium steels (0·3% C, 1% Cr), water-quenched from 1,000° C.

Fig. 21—Steel *A* 6·0 (0·08% C, 6·08% Ti), heated for ½ hr. at 700° C. and water-quenched. White rounded particles of Fe_2Ti in grain boundary and interior × 250

FIG. 27—X-ray photograph of water-quenched specimen; much lattice strain

FIG. 28—X-ray photograph of same specimen as in Fig. 27, after tempering; lattice strain relieved

FIGS. 26–28—Steel *C* 0·0 (0·38% C Ti nil

FIG. 30—X-ray photograph of water-quenched specimen; much lattice strain

FIG. 31—X-ray photograph of same specimen as in Fig. 30, after tempering; lattice strain largely maintained

FIGS. 29–31—Steel *C* 2·0 (0·41% C, 2·3% Ti)

FIG. 26—Microstructure after tempering, showing general precipitation of cementite × 2000

FIG. 29—Microstructure after tempering, showing no general precipitation of cementite, and particles of titanium nitride-carbide × 2000

FIGS. 26–31—Effect of titanium in preventing the precipitation of cementite and maintaining lattice strain during the tempering of quenched steel

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temperature, whereas with the titanium-free steel *D* 0·0, Fig. 18 shows that variation in temperature of treatment has practically no effect. The increase with 0·5% of titanium is greater than with 2%. The temperature effect is also clear in Figs. 14–17, which likewise show that small titanium additions increase hardenability considerably if a high temperature of treatment be employed. The maximum hardenability does not occur at the 4 : 1 titanium/carbon ratio but at about 1 : 1. This is due, it is suggested, to the operation of opposing factors, as the titanium content increases (*a*) the formation of TiC which, when dissolved, should increase hardenability for the same reason that it increases martensite stability and, (*b*), the reduction in " free " carbon content and grain-size and the provision of nuclei in the form of titanium nitride, undissolved carbide, and possibly aluminium compounds, which will all decrease hardenability.

A part of the increased hardenability apparently due to titanium is presumably due to aluminium, which is itself a powerful hardener. Allowing for this from Grossman's data[10] and making an approximate allowance for grain-size from his results and the McQuaid-Ehn tests referred to earlier, it appears that when the treatment temperature is 1235° C. titanium increases hardenability by about the amount that Grossman gives for an equal weight-percentage of chromium.

Corrosion Tests

Specimens were exposed to a relatively unpolluted outdoor atmosphere (south side of Swansea). The initial tendency to rust over the first three weeks was slightly reduced by amounts of titanium over 2% but the loss in weight after one year after removing the corrosion product was unaffected by the presence of titanium.

II—Titanium in Low-Alloy Steels

Four series of low-alloy steels were prepared by the United Steel Co., Ltd., in the same way as the plain carbon steels, the titanium contents varying from 0 to 5% in four steps. The alloy base compositions chosen were chromium–molybdenum, manganese–molybdenum, 1% chromium, and 3% nickel, all to contain 0·35% of carbon. Analyses, made on the portions of the bars received, are given in Table XI. The analysis figures for the alloying elements are in satisfactory agreement with those intended, and the carbon contents, which range from 0·31 to 0·40%, are also reasonably close to the intended figure, although there is a tendency for the carbon content to increase with increasing titanium content. The sulphur and phosphorus contents of only one steel in each series were determined and all are below 0·02%. The silicon and manganese contents, apart from the manganese contents of the manganese–molybdenum series, are similar to those of the plain carbon steels. As in the plain carbon steels, however, the aluminium contents are seriously high, again being about one-quarter of the titanium contents.

Microstructures of Alloy Series

The microstructures of specimens of all the steels were examined in the as-received condition and some also after quenching.

In the as-received condition the pearlite content, expressed as the percentage of surface area occupied by pearlite, decreased for all series from about 45% in the absence of titanium to about 10% with 1¼% of titanium, and pearlite was absent from the specimen containing 5% of titanium, which instead contained considerable Fe_2Ti. Grey titanium carbide crystals were found in all the titanium-bearing compositions.

After heating for 1 hr. in air at 1000° C. or 1235° C. and water-quenching, titanium-free specimens had a larger grain-size (*i.e.*, presumably the austenite grain-size) and a coarser and more acicular structure than corresponding specimens containing 1¼% of titanium ; this difference, which is shown in Figs. 24 and 25, was more marked for the chromium and nickel than for the chromium–molybdenum and manganese–molybdenum base compositions. After either heat-treatment the specimens with 5% of titanium contained no Fe_2Ti, which in these steels therefore dissolves at 1000° C. in 1 hr. Repeat tests showed that no precipitation of Fe_2Ti occurred during oil-quenching from 1000° C.

Effect of Tempering the Alloy Series

All the steels except those containing 0·25% of titanium were heated for 1 hr. in air at 1000° C. or 1235° C., quenched, tempered for 1 hr. at successively higher temperatures, and the hardness measured ; results are given in Table XII.

The response to tempering is similar to that of the plain carbon steels. With the titanium/carbon ratio of about 4 : 1 provided by 1¼% of titanium, hardness is maintained much better after quenching from 1235° C. than in the absence of titanium, to such an extent that, despite the lower as-quenched hardness of the titanium-bearing specimens, their hardness after tempering at 600° C. is considerably higher than that of the titanium-free specimens. This applies for all four alloy base compositions but to a different degree for each. The effect is most marked for the 1% chromium base compositions, for which the results are plotted in Fig. 6, and least for the 3% nickel. The former shows a stronger effect than the chromium–molybdenum base composition, although possessing only the same chromium content. It also shows a stronger effect than any

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TABLE XII—*Influence of Titanium on the Effect of Tempering Quenched Alloy Steels*

(a) *Softening of Martensite*

Mark	V.D.H., As-Quenched from 1000° C.*					V.D.H., As-Quenched from 1235° C.*				
	As-Quenched	Tempered 1 hr. at :			Percentage Drop at 600° C.	As-Quenched	Tempered 1 hr. at :			Percentage Drop at 600° C.
		200° C.	400° C.	600° C.			200° C.	400° C.	600° C.	
CM 0·0	613	568	450	316	48	606	543	456	275	53
CM 1·2	470	475	441	342	27	481	489	439	421	12
MM 0·0	615	555	438	287	49	597	531	432	249	58
MM 1·2	548	518	427	321	41	570	510	436	384	32
CR 0·0	578	538	443	275	52	588	525	439	254	57
CR 1·2	341	344	333	281	17	430	440	424	407	5
NI 0·0	601	540	389	236	61	553	509	354	197	64
NI 1·2	504	491	387	265	47	518	466	394	322	38

(b) *Precipitation-Hardening by Fe_2Ti*

Mark	Quenching Temperature	V.D.H., As-Quenched	V.D.H. when Tempered 1 hr. at			
			500° C.	600° C.	700° C.	800° C.
CM 5·0	W.Q. 1235° C.	214	226	218	241	232
	W.Q. 1000° C.	212	214	220	250	238
	O.Q. 1000° C.	205	211	209	229	225
MM 5·0	W.Q. 1235° C.	224	233	242	277	244
	W.Q. 1000° C.	222	220	225	272	252
	O.Q. 1000° C.	206	204	219	245	239
CR 5·0	W.Q. 1235° C.	211	224	239	265	243
	W.Q. 1000° C.	209	219	217	242	225
	O.Q. 1000° C.	205	215	216	223	221
NI 5·0	W.Q. 1235° C.	246	320	531	269	263
	W.Q. 1000° C.	238	359	515	267	238
	O.Q. 1000° C.	241	358	502	260	238

* Specimens heated for 1 hr. before quenching, at the temperature indicated

of the plain carbon compositions reported in Table VII. In the absence of titanium the hardness after tempering at temperatures up to 400° C. is maintained better in the chromium–molybdenum, manganese–molybdenum, and chromium but not the nickel, base compositions than in the plain carbon steels of similar carbon contents, such as steel *C* 0·0 of Table VII. With 5% of titanium precipitation-hardening occurs in all four alloy compositions, but to a marked extent only in the 3% nickel steel *NI* 5·0 in which a greater hardening effect occurs than in the carbon steels with 6% of titanium. A point of practical value is that almost the same hardness increment is obtained on reheating the *NI* 5·0 specimen oil-quenched from 1000° C. as on re-heating the specimen water-quenched from 1235° C.

Further tests were performed on *NI* 5·0 to determine the hardness–time curves at 500°, 550°, and 600° C. The specimens for these tests were water-quenched from 1000° C. after a solution treatment of 1 hr. at that temperature. The curves are given in Fig. 11. The maximum hardness obtained was 610, after 100 hr. at 500° C. At 600° C. the rate of hardening is about fifty times as great, and the peak hardness of 550 was reached in 2 hr. Comparison of Fig. 10 (plain carbon steel) and Fig. 11 (nickel steel) shows that with 5% of titanium at 600° C. 3% of nickel raises the hardness increment by 200 units and accelerates ageing about fifty times.

Tensile Tests at Elevated Temperatures on Alloy Series

Elevated-temperature tensile tests were performed on steels *CR* 1·2 and *NI* 5·0, which show

the maximum retention of hardness on tempering and the maximum precipitation-hardening respectively. Specimens were solution-treated for 1 hr. in air at, and water-quenched from, 1235° C. in the case of the *CR* 1·2 specimens and 1000° C. in the case of the *NI* 5·0 specimens, and tempered for 1 hr. at 600° C. The strengths are plotted against temperature in Fig. 12. Steel *CR* 1·2 maintains its strength up to 550° C. better than the alloy-free retentive steels but deteriorates more rapidly between 550° and 650° C. Steel *NI* 5·0 has the highest strength between 450° and 650° C. of any steel tested and maintains its strength particularly well up to 550° C., which is approximately the optimum temperature for precipitation-hardening, but the room-temperature strength is anomalously low.

Hardenability Tests on Alloy Series

Jominy hardenability tests were carried out on the four titanium-free steels and the four containing 1¼% of titanium, one set of specimens being quenched from 1000° C. and another from 1235° C. Additional specimens of the retentive steel *CR* 1·2 and the titanium-free companion composition *CR* 0·0 were treated at 900°, 1100°, and 1300° C. The experimental procedure was the same as that described in a previous section.

The results are shown diagrammatically in Fig. 34. Of the titanium-free compositions, the manganese–molybdenum has the highest hardenability, followed by the chromium–molybdenum steel, and the nickel steels have least. For all

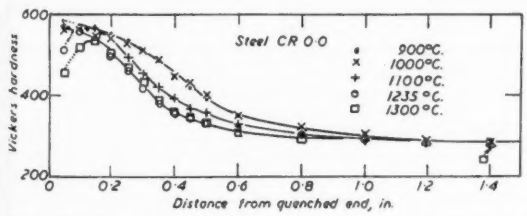

FIG. 32—Jominy hardness curves for steel *CR* 0·0 (0·32% C, Ti nil). Bars quenched from temperatures indicated

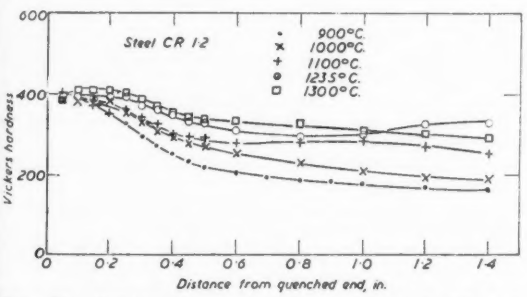

FIG. 33—Jominy hardness curves for steel *CR* 1·2 (0·32% C, 1·18% Ti). Bars quenched from temperatures indicated

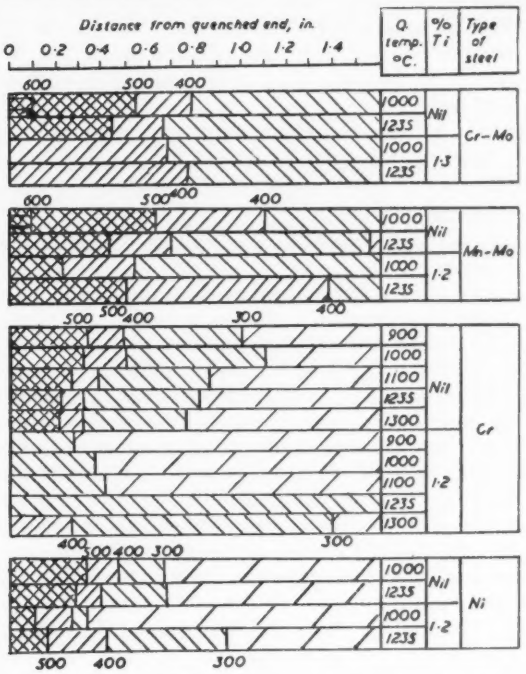

FIG. 34—Influence of titanium content and alloy composition on hardenability, as shown by the hardness distribution along Jominy bars

four compositions, the effect of raising the quenching temperature is to reduce hardenability; this can be seen from Fig. 34. Titanium (*a*) lowers the hardness at the quenched end and (*b*) maintains or raises the hardness some distance away, thus producing a flatter hardness curve. From the standpoint of improving hardenability (*b*) is desirable but (*a*) undesirable. Raising the quenching temperature accentuates (*b*) except for the chromium–molybdenum base composition, which in any case has good hardenability when quenched from 1000° C., but does not affect (*a*) much. The maximum of effect (*b*) and the minimum of effect (*a*) occur with the manganese–molybdenum base composition quenched from 1235° C. and the chromium–molybdenum base composition quenched from either temperature, and these compositions therefore give the best combination of high overall hardness and flatness of curve. Both effects are small with the nickel base composition. The complete hardness curves for the chromium base compositions *CR* 0·0 and *CR* 1·2 quenched from the various temperatures are given in Figs. 32 and 33, and for the titanium-bearing composition show the improvement in (*b*) and therefore in hardenability as the quenching temperature is raised, and the drop in hardness at the quenched end caused by titanium. The rise of the 1235° C. curve from 1 in. onwards is probably due to some precipitation-hardening by

TiC during cooling. The hardness curves for the titanium-free composition clearly show the decrease in hardenability with rise in quenching temperature.

Discussion

Within the range of compositions studied the effect of titanium on steels is a resultant of three separate effects :

(*a*) The strong affinity between titanium and carbon.

(*b*) The displacement of the γ region to higher carbon contents.

(*c*) The precipitation-hardening type of solid-solubility curve of titanium in ferrite.

From (*a*) there is derived the effect of titanium on mechanical properties in the slowly cooled state, the suppression by titanium of quench-age-hardening and embrittlement and of strain-age-hardening, and the influence of titanium in maintaining the as-quenched hardness on tempering and in improving hardenability. The assumption is made here that the effects are not due to the aluminium also present since they appear to be controlled by the titanium/carbon ratio irrespective of the carbon content, and therefore irrespective of the aluminium/oxygen ratio. The high quenching temperature necessary for the latter effects militates against their practical use. The effect on hardness is influenced by other alloy elements in an individual way that cannot be entirely predicted from their own known influence on tempered hardness. Thus chromium and molybdenum combined are known to maintain hardness on tempering better than chromium alone,[11] but the chromium–molybdenum–titanium combination is inferior to the chromium–titanium combination. The effect on hardenability is influenced by other alloy elements in the expected way, so that the base compositions, arranged in order of hardenability, have the same order whether carrying titanium or not. The conclusion drawn earlier, that titanium has about the same effect on hardenability as an equal weight percentage of chromium if a high treatment temperature be employed for the titanium steel, is borne out by comparison of the results for *D* 0·5 (0·5% of carbon, 0·7% of titanium) and *CR* 0·0 (0·3% of carbon, 1·1% of chromium) in Figs. 19 and 32.

From (*b*) there results the disappearance of the martensitic structure and hardness from quenched steels when the titanium content is high enough.

From (*c*) there results the possibility of marked precipitation-hardening. Alloy elements can modify this effect, and the 1% of aluminium present in all the precipitation-hardening alloys tested may therefore have exerted a modifying influence. Chromium, manganese, and molybdenum in the quantities examined were found to reduce the hardness increment to a small amount, whereas 3% of nickel was found to increase it considerably and to reduce the solution and tempering temperature necessary. The reduction in tempering temperature means of course that the maximum permissible service temperature is reduced. It is known[12] that silicon also increases the hardness increment caused by small amounts of titanium, and probably some other elements have the same effect. The possibility therefore exists of a combination of small additions permitting a substantial hardness increment with a relatively small titanium content.

Thus although titanium is not a generally useful alloying element there are a number of applications in which it has a beneficial effect. However, there may be considerable difficulties in large-scale manufacture in keeping down the nitride and aluminium contents.

Acknowledgment

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The Embrittlement of Chromel and Alumel Thermocouple Wires*

By W. I. Pumphrey, M.Sc.†

SYNOPSIS

An investigation has been made of the surface embrittlement that occurs in Chromel–Alumel thermocouples heated to temperatures above 800° C. in uncleaned stainless-steel protecting sheaths. Under such conditions and also under purely oxidizing conditions Alumel wire suffers more severe surface attack and exhibits more pronounced embrittlement than Chromel. Embrittlement in the stainless-steel sheath does not occur when the sheath is baked for 3 hr. at 800° C. before use, but reappears if the baked sheath is oiled before testing. It is suggested that oil or other carbonaceous or sulphur-bearing matter remaining in the protecting tube after manufacture is the cause of the embrittlement, and that all stainless-steel sheaths for use with Chromel–Alumel thermocouples should be thoroughly cleaned before use.

Introduction

It was reported in a previous paper[1] that a Chromel–Alumel thermocouple used for determining cooling rates at the centre of a silver cylinder during quenching from 825° C. was rapidly attacked by the silver. Subsequent experience, however, suggested that the embrittlement of the Chromel and Alumel wires was due, not to attack by the silver, but to conditions inside the stainless-steel sheath within which the thermocouple was contained. This suggestion was also supported by the fact that embrittlement was not observed in unsheathed thermocouples used under similar conditions.

Subsequent to the publication of the "Symposium on the Contamination of Platinum Thermocouples"[2] it was felt that it would be of value to carry out a limited investigation of the embrittlement of Chromel and Alumel thermocouple wires and it was thought that publication of the results of this investigation might be of interest to other users of this type of base-metal thermocouple.

Results

As a preliminary to the investigation a new Chromel–Alumel thermocouple was made up from 22 S.W.G. wire and fitted inside a stainless-steel sheath made from a new and uncleaned length of welding-quality 18/8 stainless-steel tubing sealed at one end by welding. Thermocouple and sheath were placed in a furnace and maintained for 2 hr. at a temperature of 800°–825° C. in a mildly oxidizing atmosphere. On removal from the tube after this treatment the couple was re-calibrated and was found to read some 40° C. low at a temperature of 659° C. Subsequent examination of both wires revealed that while the Chromel wire was free from cracks and was still ductile, the Alumel wire had suffered severe surface attack and consisted of a hard and brittle outer skin surrounding a ductile core. The appearance of the two wires under the microscope is shown in Figs. 1, 2, and 3. It will be seen that although the Chromel wire was not entirely free from surface attack the degree of attack was considerably less than in the case of the Alumel wire.

The surface cracks observed in the Alumel wire were, in the main, intercrystalline (*see* Fig. 2) and were similar to those observed in platinum and platinum–rhodium thermocouple wires by Reeve and Howard.[3]

In addition to the above effect it was observed that in the contaminated region of the thermocouple, the Chromel wire, which is normally not attracted, was attracted by a magnet to the same extent as the Alumel wire which was magnetic both before and after heating in the sheath.

On completion of the experiment the stainless-steel tube was baked in a furnace for 3 hr. at 800° C., a new thermocouple made up and inserted in the tube, and the baked tube and enclosed thermocouple placed in a furnace and maintained for some time at a temperature of

* Received 30th July, 1947.

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800°–825° C. Under these conditions no appreciable embrittlement of the thermocouple wires or alteration in the calibration was observed after 6 hr. at the temperature of the experiment. The baked sheath was then soaked for several hours in a mineral quenching oil, and on repetition of the test in the oiled sheath, surface embrittlement was observed in both wires of the thermocouple after 3 hr. at the temperature of the test. As will be seen from Fig. 4, after this treatment the Alumel wire was attacked to a greater extent than the Chromel but both were attacked to a lesser degree than wires treated in the tube before baking and oiling. After treatment in the oiled tube the Chromel wire remained unattracted by a magnet.

In order to determine the extent of the surface attack in Chromel and Alumel wires maintained at a high temperature for some length of time under mildly oxidizing conditions in the absence of a stainless-steel sheath, small pieces of 22 S.W.G. Chromel and Alumel wire were placed in separate, specially cleaned platinum crucibles and maintained in a furnace at a temperature of 800°–825° C. for up to 360 hr. Specimens were removed from the furnace at regular intervals for microscopic and magnetic examination. The results of this examination are shown in Figs. 5–10.

As will be seen, the mode of attack of the wires under the purely oxidizing conditions of the treatment was somewhat different from the mode of attack of the wires when in the stainless-steel sheath. In particular, no pronounced intercrystalline surface cracking was observed, a result to be expected in view of the noncorrosive conditions of the test. The Alumel wires were attacked to a considerably greater extent than the Chromel wires, with the formation of a more pronounced hard and brittle surface layer, but under the conditions of the test, even after 360 hours at 800°–825° C. the Chromel wires remained unattracted by a magnet whilst the Alumel wire was still attracted.

Conclusions

Under purely oxidizing conditions and in the absence of any corrosive effects both Alumel and Chromel wires suffer progressive surface oxidation when maintained for some time at a temperature of 800°–825° C. Under such conditions the Alumel wire is attacked to a considerably greater extent and exhibits a more pronounced brittle surface layer than the Chromel wire. When a Chromel–Alumel thermocouple is heated for a short time at a temperature of 800°–825° C. in a new but uncleaned stainless-steel sheath the Alumel wire develops pronounced surface brittleness, and after some 2 hr. at this temperature the couple may read up to 40° C. low. If the thermocouple sheath is baked for 3 hr. at 800° C. before the test the embrittlement of the thermocouple is not observed ; oiling of the baked sheath before testing, however, causes a recurrence of the embrittling action. After the treatment in the unbaked tube the Chromel wire which is normally not attracted by a magnet, is attracted to the same extent as the Alumel wire. This alteration in the magnetic properties of the Chromel wire is not observed after 360 hr. under mildly oxidizing but noncorrosive conditions at a temperature of 800–825° C.

The baking of the stainless-steel sheath would have the effect of burning off oil and carbonaceous matter remaining in the tube after manufacture, and from the above results it would thus seem that whilst Chromel and Alumel wires both suffer surface oxidation to some extent at temperatures of 800° C. and above, with consequent surface embrittlement, this embrittlement is considerably aggravated and hastened if the materials are heated in the presence of oil or other carbonaceous or sulphur-bearing matter. Alumel, being much more subject to oxidation than Chromel, is attacked to a greater extent at high temperatures in the presence of carbonaceous or sulphur-bearing matter, and under such corrosive conditions the surface attack of the Alumel wire may result in the occurrence of intercrystalline surface cracking. In view of the results obtained with the baked stainless-steel sheath it seems not unreasonable to assume that the material of the thermocouple sheath itself contributes in no way to the embrittlement of the contained thermocouple wires and that the corrosive attack at high temperatures with the unbaked tube is due entirely to foreign matter not removed from the tube. For this reason all stainless-steel sheaths for use with Chromel–Alumel thermocouples should be thoroughly cleaned before use, and such thermocouples should not be used unprotected under conditions in which there is likelihood of high-temperature attack by oil or other sulphur-bearing matter.

The reason for the alteration in the magnetic properties of Chromel wire after high-temperature treatment in the unbaked tube is not known, particularly since the effect appears to be absent when the material undergoes oxidation attack or attack in the presence of a known oil.

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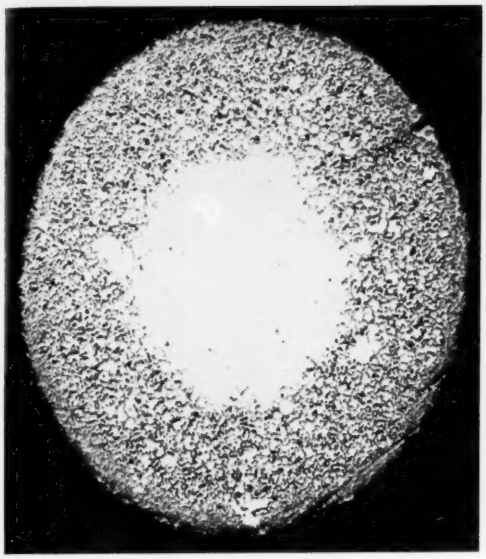

FIG. 1—Alumel wire of embrittled thermocouple, unetched. Wire attracted by magnet. × 100

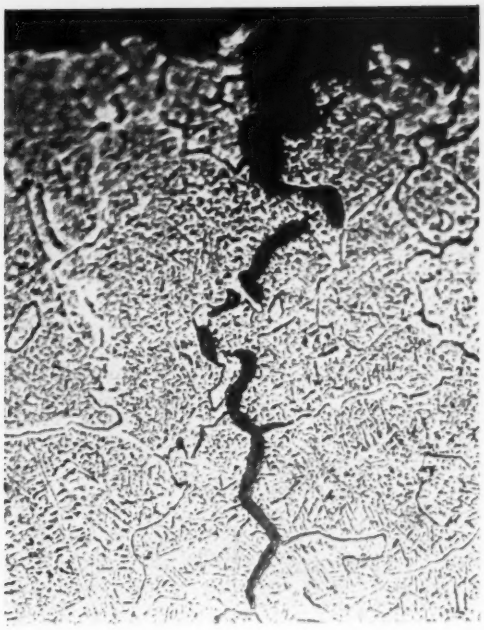

FIG. 2—Surface cracks in the Alumel wire shown in Fig. 1, unetched. × 500

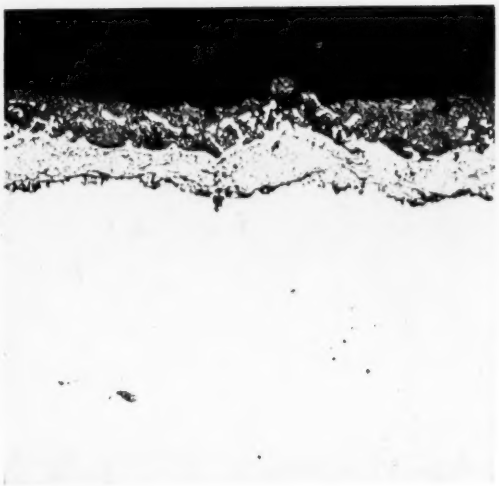

FIG. 3—Chromel wire of embrittled thermocouple, unetched. Wire attracted by magnet. × 500

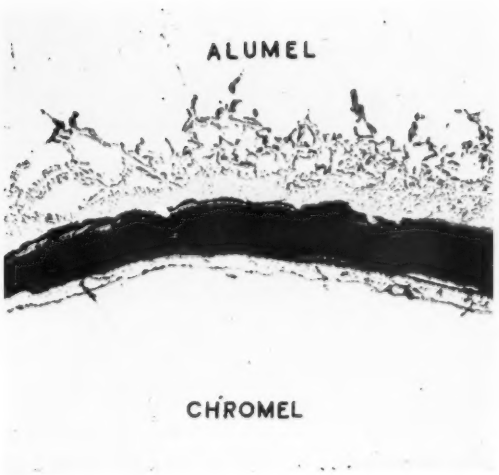

FIG. 4—Chromel and Alumel wires after 3 hr. at 800°-825° C. in oiled stainless-steel sheath, unetched. × 250

FIG. 5—Chromel wire before test, unetched. Wire not attracted by magnet × 500

FIG. 6—Chromel wire after 180 hr. at 800°-825° C., unetched. Wire not attracted by magnet × 500

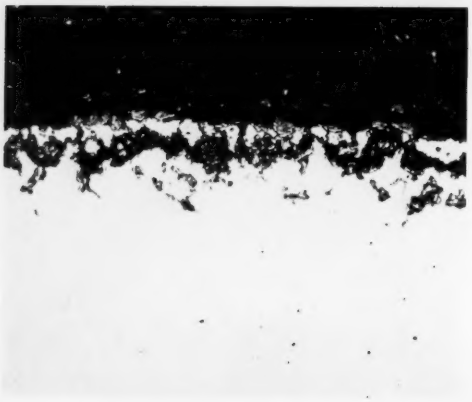

FIG. 7—Chromel wire after 360 hr. at 800°-825° C., unetched. Wire not attracted by magnet × 500

FIG. 8—Alumel wire before test, unetched. Wire attracted by magnet × 500

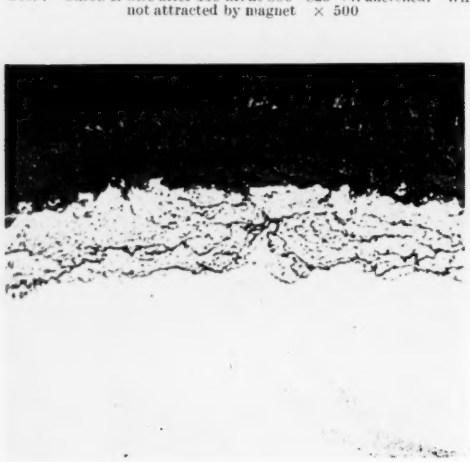

FIG. 9—Alumel wire after 180 hr. at 800°-825° C., unetched. Wire attracted by magnet × 500

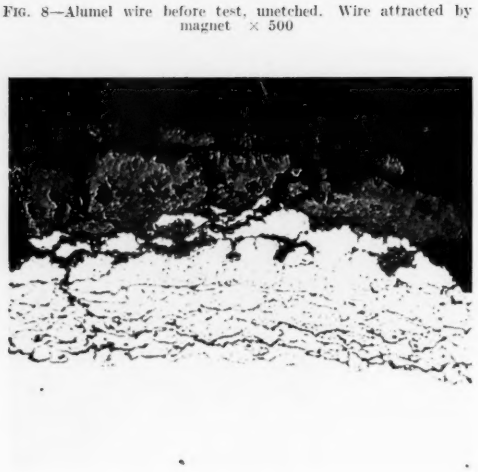

FIG. 10—Alumel wire after 360 hr. at 800°-825° C., unetched. Wire attracted by magnet × 500

Influence of Carbon on the Equilibria between Steel and Slag in the Acid Open-Hearth Furnace*

By Dr. P. Herasymenko†

SYNOPSIS

Contrary to the conclusions of certain previous investigators, the opinion is expressed that equilibrium can easily be obtained for most of the slag–metal reactions taking place in acid open-hearth heats, and details are given of four such heats in which equilibrium between the slag and the metal was attained and maintained. It is only in the case of certain manufacturing methods that the bulk of the slag does not take part in the exchange reactions. The reactions $Mn + FeO = MnO + Fe$, *and* $Si + 2MnO = SiO_2 + 2Mn$, *and also the mechanism of the oxidation of carbon in the acid open-hearth furnace, are discussed, and the author's conclusions are contrasted with those of other workers.*

THE workers of the Kaiser-Wilhelm-Institut für Eisenforschung, on the basis of their investigations of the principal reactions in steelmaking, have come to the conclusion that equilibria between slag and metal in acid open-hearth practice are not, and cannot be, attained, owing to the peculiarities accompanying the oxidation of carbon.[1, 2] It is the intention of the present author, however, to show in this paper that in the case of acid open-hearth-furnace heats the equilibria can easily be reached for most reactions. It is only with certain manufacturing methods that the acid slags become so persistently viscous that the bulk of the slag does not take part in exchange reactions between the metal bath and the film of slag adjacent to the steel.

The conditions of equilibria in the pure Fe–Mn–Si–O system saturated with solid silica were investigated by Körber and Oelsen,[3] who found that the equilibria for the reactions:

$$\begin{aligned} Mn + FeO &= MnO + Fe, \\ Si + 2FeO &= SiO_2 + 2Fe, \\ Si + 2MnO &= SiO_2 + 2Mn, \\ Fe + O &= FeO, \end{aligned}$$

can be expressed by the equations:‡

$$k^{Fe}_{Mn} = \frac{[Mn](FeO)}{(MnO)} \quad \ldots\ldots(1)$$

$$k^{Fe}_{Si} = [Si](FeO)^2 \quad \ldots\ldots(2)$$

$$k^{Mn}_{Si} = \frac{[Si](MnO)^2}{[Mn]^2} \quad \ldots\ldots(3)$$

$$k^{O}_{Fe} = \frac{[O]}{(FeO)} \quad \ldots\ldots(4)$$

The expressions (1) to (4) are practically constant (at not too high contents of manganese). The k values in these equations can be regarded as equilibrium constants, the numerical values of which are given in Table I. Both the slag and the metal phases seem to obey the law of ideal solutions.

The pure Fe–Mn–Si–O system saturated with SiO_2 is a bivariant one, *i.e.*, the composition of both liquid phases is determined by the concentration of one component in one liquid phase and by the temperature. Thus, the curves in Fig. 1 represent the dependence of (MnO), [Si], and [O] on the manganese content in liquid steel. If the content of (FeO + MnO) in the slag is lowered by additions of "inert" oxides, such as CaO or Al_2O_3, the activity coefficients of the main constituents of the slag remain unchanged, so that the content of (MnO), [Si], and (O) can be calculated from the equilibrium constants valid for the pure system. Dilution of (FeO + MnO) in the slag leads to an increase of [Si] and to a decrease of [O] in liquid steel, as can be seen by comparing the curves for 50%, 40%, and 30% of (FeO + MnO) at 1600° C. (Fig. 1).

According to the work of Körber[4] and Oelsen,[5] the presence of small amounts of chromium and nickel (up to 5%) in liquid steel has no influence on the activity coefficients of the principal constituents of the steel or slag.

Experimental Heats

The author can instance many acid heats in which an equilibrium between slag and metal was establishing spontaneously throughout the whole steel-refining process. Four heats, the data for which are given in Table II, may serve as examples. These heats were made in a 50-ton

‡ The convention whereby [] indicates concentrations in the metal phase and () indicates those in the slag phase is used throughout this paper.

* Received 25th April, 1947.

† Munich, Bavaria.

TABLE I—*Values of the Equilibrium Constants for the Pure Fe–Mn–Si–O System*

Temp., ° C.	$k^{Fe}_{Mn} = \frac{[Mn](FeO)}{(MnO)}$	$k^{Fe}_{Si} = [Si](FeO)^2$	$k^{Mn}_{Si} = \frac{[Si](MnO)^2}{[Mn]}$	$k^{O}_{Fe} = \frac{[O]}{FeO} \times 10^3$	$k^{O}_{Mn} = \frac{[Mn][O]}{(MnO)} \times 10^4$	$k^{O}_{Si} = [Si][O]^2 \times 10^5$
1500	0·049	2·25	925	1·3	0·64	0·38
1550	0·065	4·4	1030	1·71	1·1	1·4
1600	0·085	8·4	1150	2·13	1·8	3·8
1650	0·110	15·3	1280	2·55	2·8	9·9
1700	0·141	27·4	1395	2·97	4·2	24·2

furnace, the charge consisting of either (*a*) liquid steel premelted from a high-quality pig iron and scrap in a basic open-hearth furnace, or (*b*) of 33% of solid scrap, preheated in the acid furnace, and 60% of liquid steel, premelted in the basic furnace. Previous to the charging of the steel, approximately 600–800 kg. of the acid slag from previous heats of approximately the same composition was placed on the furnace bottom and heated, until it melted (case (*a*)) or until it was charged with the solid scrap (case (*b*)). The required content of alloying elements (nickel, chromium, and molybdenum) was adjusted in the early stages of the heats ; no other additions to the slag or metal were made during the whole process of steel refining. The temperature was measured by a " Pyropto " optical pyrometer, on the surface of liquid steel in the spoon, after the removal of slag. The corrected temperatures given in Table II can be regarded as true temperatures within the range of ± 15° C.; these values, therefore, can serve only as comparatively rough indications of the bath temperature. The courses of two of the heats (heats *E* and *G*) are diagrammatically shown in Figs. 2 and 3, respectively.

The oxygen and nitrogen contents of the steel were determined by the vacuum-fusion method, on samples which solidified without the evolution of gas. Determination of the oxygen content was

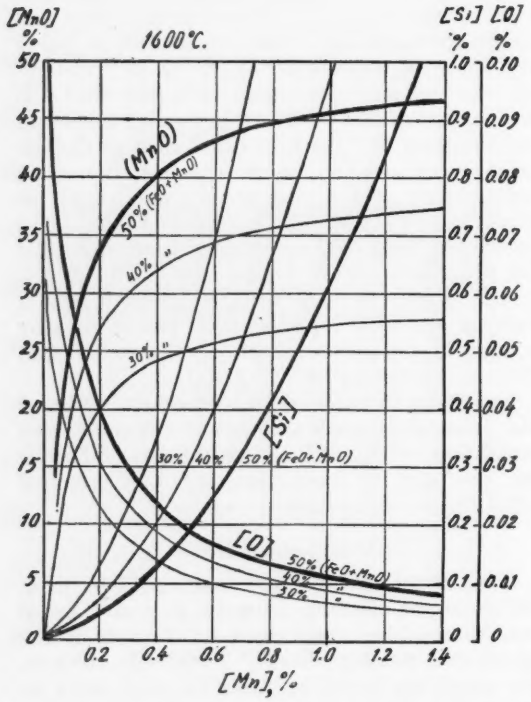

FIG. 1—Equilibrium composition of metal and slag in the pure Fe–Mn–Si–O system at 1600° C. and at 50%, 40%, and 30% of (FeO + MnO), according to Körber and Oelsen

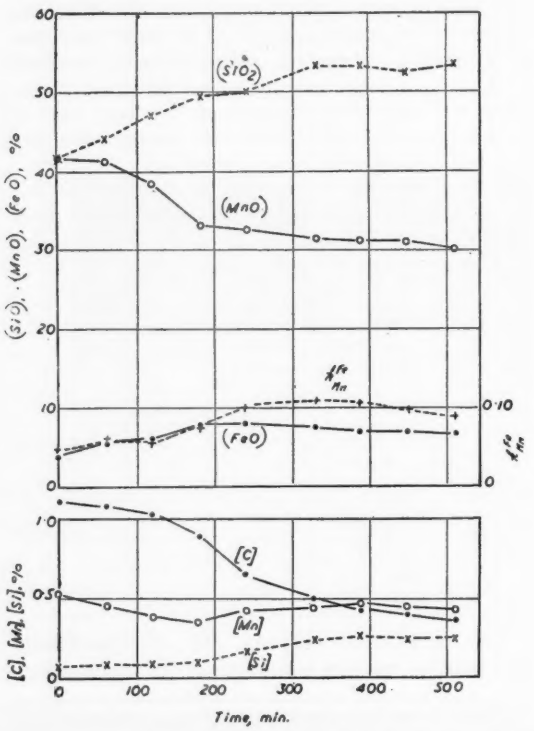

FIG. 2—Data from heat *E*

TABLE II—*Data for Four Experimental Heats*

Heat No.	Time, min.	Temp., °C.	Slag							Steel						
			(FeO), %	(MnO), %	(Cr), %	(Al), %	(V), %	(Ti), %	(SiO_2), %	[C], %	[Mn], %	[Cr], %	[Si], %	[V], %	[O], %	[N], %
*E*1	0	1500	3·9	41·6	4·30	1·02	0·63	0·13	41·8	1·11	0·52	1·21	0·06	0·12	...	...
*E*2	60	1520	5·6	41·4	4·87	0·99	0·58	...	44·2	1·08	0·44	1·24	0·07	0·11	...	...
*E*3	120	1540	5·8	38·4	...	1·02	0·50	...	47·1	1·03	0·38	1·24	0·06	0·11	...	...
*E*4	180	1560	7·8	33·2	4·09	0·74	0·46	...	49·7	0·88	0·33	1·24	0·08	0·11	...	...
*E*5	240	1600	7·9	32·3	3·26	0·72	0·42	0·11	50·1	0·63	0·41	1·25	0·15	0·11	...	...
*E*6	330	1620	7·6	31·6	3·05	0·72	0·40	...	53·4	0·49	0·45	1·25	0·24	0·12	...	...
*E*7	390	1620	7·1	31·1	2·88	0·67	0·39	...	53·3	0·43	0·46	1·24	0·25	0·12	...	...
*E*8	450	1620	7·0	31·1	2·88	0·65	0·38	...	52·5	0·39	0·44	1·25	0·25	0·11	...	...
*E*9	510	1620	6·5	30·0	...	0·62	0·37	...	53·6	0·37	0·42	1·73	0·23	0·11	...	...
*F*1	0	...	4·4	39·2	3·5	1·7	0·6	0·20	45·0	1·15	0·56	1·32	0·09	0·14	...	...
*F*2	60	1530	5·6	40·1	2·95	1·48	0·56	...	47·0	1·11	0·52	1·33	0·08	0·14	0·0030	0·004
*F*3	120	1570	6·0	38·9	3·1	1·48	0·53	...	48·0	1·05	0·46	1·32	0·08	0·14	0·0065	0·004
*F*4	180	1590	8·8	36·0	3·6	1·32	0·49	0·19	50·0	0·97	0·41	1·32	0·11	0·13	0·0070	0·005
*F*5	240	1610	8·1	33·6	2·8	1·32	0·40	...	50·5	0·82	0·45	1·34	0·19	0·14	0·0076	0·004
*F*6	300	1615	7·8	31·9	2·7	1·27	0·33	...	53·5	0·67	0·47	1·34	0·24	0·13	0·0080	0·004
*F*7	360	1620	7·7	29·2	2·6	1·22	0·30	0·20	55·0	0·54	0·48	1·33	0·31	0·13	...	...
*F*8	420	1620	6·0	27·5	2·6	1·16	0·28	...	55·0	0·45	0·47	1·35	0·32	0·13	0·0085	0·003
*F*9	480	1615	6·2	26·5	2·6	1·16	0·28	0·20	55·0	0·39	0·45	1·33	0·33	0·13	0·0079	0·002
*F*10	510	1610	6·0	26·8	2·6	1·11	0·29	...	...	0·36	0·44	1·32	0·32	0·13	0·0090	0·002
*G*1	0	1515	6·4	38·6	3·10	1·16	0·51	0·18	45·5	1·02	0·63	1·11	0·07	0·10	0·0030	0·005
*G*2	45	1530	7·2	37·5	2·80	1·06	0·46	...	46·0	0·99	0·47	1·12	0·07	0·11	0·0050	0·004
*G*3	105	1555	7·7	36·7	2·74	0·95	0 39	...	48·5	0·95	0·43	1·13	0·09	0·11	0·0061	0·003
*G*4	165	1590	7·7	35·9	2·46	0·95	0·33	0·14	51·5	0·76	0·48	1·12	0·18	0·11	0·0073	0·004
*G*5	225	1620	6·1	33·0	2·40	0·90	0·31	...	53·5	0·62	0·48	1·13	0·26	0·11	0·0076	0·004
*G*6	285	1635	5·9	32·4	2·32	0·90	0·30	...	54·5	0·43	0·51	1·14	0·33	0·11	0·0084	0·003
*G*7	345	1635	6·1	31·3	2·35	0·90	0·29	0·15	54·2	0·38	0·51	1·14	0·33	0·11	0·0089	0·003
*H*1	0	1480	5·9	40·5	1·58	1·20	...	...	47·1	1·14	0·62	0·68	0·07	0·21	0·0046	0·005
*H*2	60	1545	5·9	40·5	1·70	1·31	...	...	47·9	1·02	0·56	0·68	0·08	0·20	0·0037	0·005
*H*3	120	1575	7·8	37·2	1·80	1·12	0·66	...	48·0	0·96	0·45	0·67	0·11	0·19	0·0054	0·004
*H*4	180	1610	7·3	35·3	1·64	1·16	0·67	...	51·8	0·76	0·45	0·67	0·18	0·19	0·0069	0·003
*H*5	240	1605	7·5	32·6	1·62	1·05	0·61	...	53·0	0·52	0·46	0·69	0·21	0·20	...	...
*H*6	300	1610	7·3	31·3	1·54	1·04	0·59	...	54·2	0 40	0·46	0·68	0·24	0·21	0·0083	0·002
*H*7	360	1620	7·3	30·8	1·55	0·98	0·53	...	55·6	0·39	0·47	0·68	0·27	0·21	0·0103	0·002
*H*8	390	1615	7·1	30·8	1·52	1·00	...	...	57·2	0·38	0·48	0·67	0·28	0·22	0·0105	0·003
*H*9	420	1615	6·7	30·7	1·52	0·99	...	...	...	0·37	0·48	0·68	0·25	0·22	0·0097	0·002

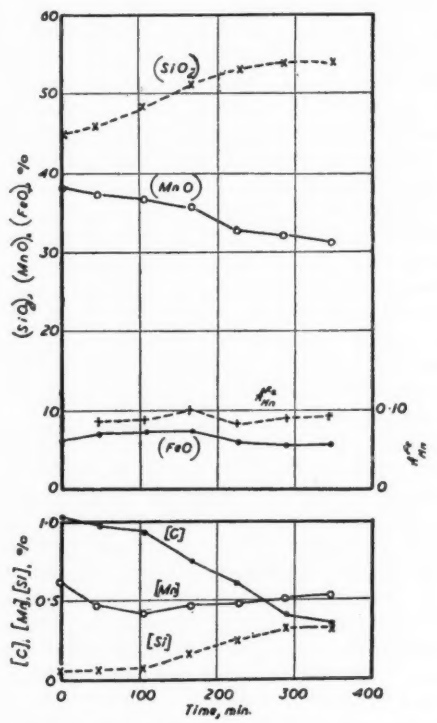

FIG. 3—Data from heat *G*

repeated several times for each sample, and the separate values differed by only ± 0·0004% from the mean values given in Table II. Some additional data on the oxygen content are given in Table III, where the analyses of the two acid heats are tabulated.

Reaction $Mn + FeO = MnO + Fe$

The values of $k^{Fe}_{Mn} = [Mn]\,(FeO)/[MnO]$ calculated from the data of Table II are plotted against temperature in Fig. 4. In the same diagram the

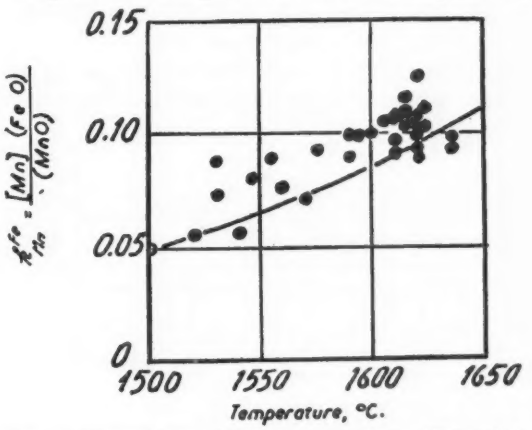

FIG. 4—Dependence of k^{Fe}_{Mn} on temperature. The points are taken from Table II, and the curve is drawn according to the data of Körber and Oelsen

TABLE III—*Analyses for Two Experimental Acid Heats*
Mean temperature 1625° C. (±15°)

Sample No.	Time, min.	Slag			Steel			
		(SiO_2), %	(FeO), %	(MnO), %	[C], %	[Mn], %	[Si], %	[O], %
				Heat K				
4	165	49·2	6·7	36·0	0·64	0·47	0·21	0·0090
5	195	52·5	6·7	32·4	0·54	0·49	0·28	0·0087
6	225	54·3	6·8	30·8	0·48	0·50	0·32	0·0090
7	255	55·6	6·8	30·6	0·41	0·52	0·34	0·0090
8	285	55·8	7·2	30·4	0·37	0·52	0·34	0·0095
9	315	56·1	7·8	30·1	0·34	0·49	0·32	0·0104
				Heat M				
2	60	...	...	...	1·00	0·58	0·12	0·0065
3	120	48·2	7·6	39·0	0·97	0·52	0·13	0·0069
4	180	51·3	7·8	38·0	0·76	0·52	0·17	0·0092
5	240	53·3	7·7	35·2	0·57	0·52	0·22	0·0100
6	275	56·4	7·8	31·1	0·49	0·52	0·30	0·0100
7	300	57·2	8·0	29·5	0·41	0·52	0·34	0·0090

course of k_{Mn}^{Fe} as found by Körber and Oelsen is drawn as a solid line. On the average, the present author's values for k_{Mn}^{Fe} lie slightly above the equilibrium curve of Körber and Oelsen, and this can be attributed partly to the rough determination of the temperatures and partly to the inhomogeneous distribution of the FeO content in the slag, which will be discussed later. In any case the deviations of the author's values for k_{Mn}^{Fe} from the equilibrium curve are insignificant and lie within the range of experimental error. It may be noted that the scattering of the separate k_{Mn}^{Fe} values in the laboratory work of Körber and Oelsen (which led to the establishment of the equilibrium curve) was even greater than in Fig. 4. The values for k_{Mn}^{Fe} are plotted against the carbon content of the steel in Fig. 5, using the present author's data as well as those of Körber and Oelsen. The results show that the reaction

$$Mn + FeO = MnO + Fe$$

attains particularly well the equilibrium in all stages of the author's heats and is independent

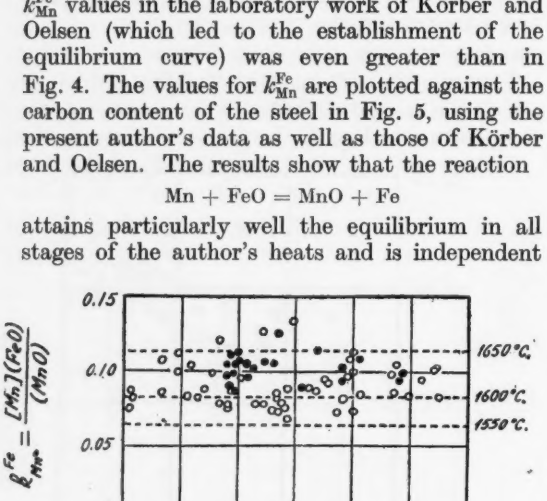

FIG. 5—Experimental values for k_{Mn}^{Fe} plotted against [C]. The solid circles represent the author's data and the clear circles the data of Körber and Oelsen at the mean temperature of 1600° C.

of the carbon content and of the rate of carbon elimination.

The mixing action of the boil produces rather favourable conditions for the acceleration of the exchange reactions between the liquid steels and slag. On the other hand, it also promotes a more rapid dissolution of silica from the walls and hearth of the furnace, so that the slag becomes approximately saturated with silica. The solubility of silica in the acid open-hearth slags which may contain considerable quantities of "inert" oxides (CaO, Al_2O_3, etc.) is higher than in the pure Fe–Mn–Si–O system. With increasing content of SiO_2 in the slag the latter becomes viscous, and the equalization of oxide concentrations in the slag is made more difficult when there is an excessive amount of slag. Small droplets of steel ejected from the bath during the boil can remain in suspension in the viscous slag and are oxidized at the gas–slag interface, thus increasing the total (analytical) content of iron oxide. Therefore the usual spoon samples of slag will not always correspond to the composition of the slag film adjacent to the metal bath which is in equilibrium with liquid steel.

An example of such an excessive increase in the ferrous oxide content is given in Table IV, in which the analyses of steel and slag in an acid heat are tabulated. The heat was made at the end of a campaign, when the furnace bottom was not in perfect condition. During the boil, parts of the bottom loosened and rose to the slag, resulting in a premature increase of the silica content of the slag as well as of the total amount of slag. About 1 hr. after the silica content had reached its highest value, a rapid increase of the analytical content of ferrous oxide in the slag took place, much above the equilibrium value.

TABLE IV—*Steel and Slag Analyses for an Experimental Acid-Furnace Heat*

Sample No.	Time, min.	Temp., ° C.	Slag			Steel		
			(SiO_2), %	(FeO), %	(MnO), %	[C], %	[Mn], %	[Si] %
1	0	1570	50·2	8·1	36·2	0·86	0·43	0·10
2	60	1590	53·3	7·8	33·7	0·67	0·46	0·18
3	120	1610	56·7	8·1	31·4	0·54	0·50	0·27
4	150	1620	57·2	9·1	30·7	0·46	0·50	0·32
5	180	1630	57·2	9·3	30·0	0·41	0·49	0·35
6	210	1630	55·8	11·5	28·2	0·38	0·50	0·37
7	240	1620	55·5	11·9	27·8	0·37	0·50	0·37
8	300	1620	53·9	14·7	25·5	0·36	0·48	0·37

Since the bath had ceased to boil at the same time, the equalization of the FeO content in the slag could not proceed fast enough. The increase of the FeO content took place in the upper layers of slag. Were the increased content of ferrous oxide equally distributed in all layers of the slag, it should have produced considerable changes in the metal bath, but this was not the case. With acid heats of the composition as exemplified by the data in Table II, such an abnormal increase in the iron oxide content occurred only in exceptional cases, like the one just described.

An easy attainment of equilibria for the reaction Mn + FeO = MnO + Fe can be proven from the data published by Malcor[6]; the course of an acid heat according to Malcor's observations is shown in Fig. 6. The curve of the values for k_{Mn}^{Fe} calculated from analytical data agrees well with the equilibrium values of Körber and Oelsen.

Bramley, Maddocks, and Tateson[7] have published a detailed description of an acid heat for which the manufacturing method differed considerably from that of the acid heats discussed above. The diagram of the heat studied by Bramley and his co-workers is given in Fig. 7. In this case the charge contained a large amount of silicon, so that the first sample after the melt-out contained 0·4% of silicon; the manganese content of the metal bath was low throughout the heat. The boil and the fluidity of the slag were maintained by frequent additions of iron ore and limestone to it. At the early stages of heat, the steel was not in equilibrium with the slag, as can be inferred from the high values for k_{Mn}^{Fe}. About 2 hr. after the melting-out—when the silicon in the bath was oxidized—the slag very quickly attained equilibrium with the metal bath, and the values for k_{Mn}^{Fe} fell within the range 0·08–0·10, which corresponds well with the equilibrium values as determined by Körber and Oelsen for steelmaking temperatures. Small, irregular variations of k_{Mn}^{Fe} might be due to analytical errors (in the determination of a low content of manganese) and to a certain inhomogeneity of the slag owing to frequent additions of ore.

The reaction CrO + Mn = Cr + MnO also attained equilibrium in the course of the author's heats. The observed values of

$$k_{Cr}^{Mn} = \frac{[Cr](MnO)}{(Cr)[Mn]}$$

agree well with the values for k_{Cr}^{Mn} derived from

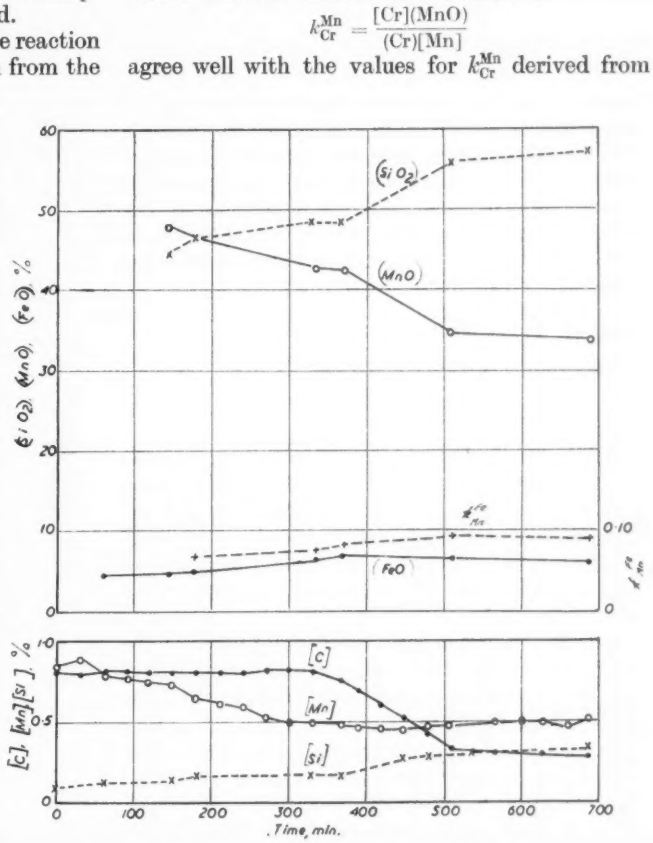

FIG. 6—Data for an acid heat observed by Malcor

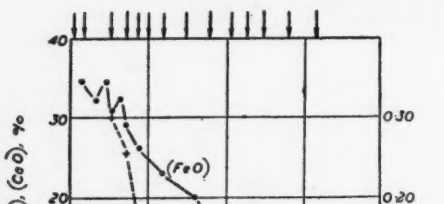

FIG. 7—Data for an acid heat observed by Bramley, Maddocks, and Tateson. The arrows indicate the times of ore and limestone additions

the laboratory measurements by Körber and Oelsen. As shown in Fig. 8, the constant k^{Mn}_{Cr} is hardly influenced by the temperature. It should be mentioned that Körber and Oelsen have shown that at high temperatures chromium in acid slags exists mainly in a divalent form, *i.e.*, as CrO. On cooling it oxidizes to trivalent chromium. In Table II the chromium content is expressed as metal, and not as oxide. The same applies for aluminium, vanadium, and titanium.

Reaction $Si + 2MnO = SiO_2 + 2Mn$

The values of the " constant "

$$k^{Mn}_{Si} = \frac{[Si](MnO)^2}{[Mn]^2},$$

derived from the author's acid open-hearth heats at low carbon contents (0·40–0·30%), also agree well with the values of k^{Mn}_{Si} as found by Körber and Oelsen[8] for the pure (carbonless) Fe–Mn–Si–O system. However, at higher contents of carbon, the value k^{Mn}_{Si} decreases regularly with carbon content.

The bath temperatures are usually lower at the beginning of the boil, *i.e.*, when the liquid steel has higher contents of carbon, than in the later stages of heat, but the low values for k^{Mn}_{Si} cannot be explained by the low temperature of the bath. According to the determinations made by Roth,[9] the heat of reaction of $Si + 2MnO = SiO_2 + 2Mn$ is fairly low (between 8,100 and 15,100 cal./mol.), so that a strong influence of temperature on the equilibrium constant of this reaction cannot be expected. Indeed, Körber and Oelsen have found that k^{Mn}_{Si} varies only slightly with temperature.

As already mentioned, other manganese reactions attained equilibrium in the course of the author's heats, and there is no reason to assume that the case of the reaction $Si + 2MnO = SiO_2 + 2Mn$ should be specifically different. The only reasonable explanation of the lowering of the k^{Mn}_{Si} values with increasing carbon content is that this effect is due to the influence of carbon on the activity coefficient of some of the reaction partners. Indeed, Körber and Oelsen have found that the relative activity coefficient of silicon in liquid steel increases at high carbon contents, so that a steel with high content of carbon can contain a lower content of silicon than a carbonless steel in equilibrium with slag of the same composition. This influence of carbon on the activity of silicon was not investigated in detail by Körber and Oelsen at moderate contents of carbon ; these authors limited themselves to a statement that the variation in the activity coefficient of silicon is perceptible only at carbon contents higher than 1%, although an analysis of their data shows that the effect of carbon becomes already evident at lower carbon concentrations.

When plotting the values of k^{Mn}_{Si} against the carbon content from our data as well as from those calculated from the laboratory experiments of Körber and Oelsen, the present author obtained

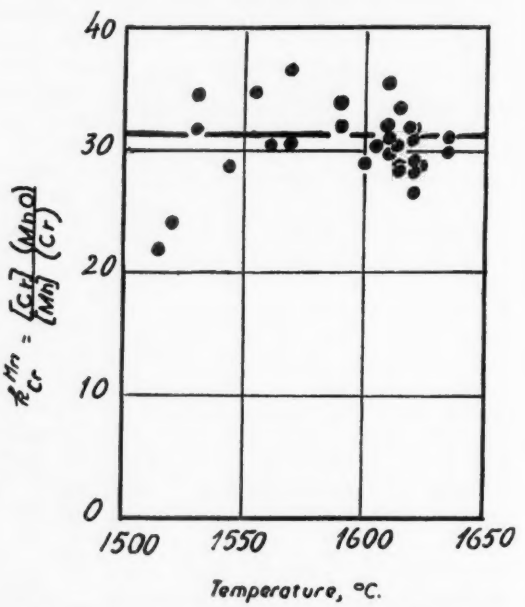

FIG. 8—Independency of the constant k^{Mn}_{Cr} on the temperature

the result illustrated in Fig. 9, which shows that the value of k_{Si}^{Mn} rapidly decreases in the interval between 0·4 and 0·9% of carbon. The solid line represents a curve of average values.

The decrease of the k_{Si}^{Mn} value, and thus the decrease in silicon with increasing carbon content, points to a relative increase of the activity coefficient of silicon with respect to other elements in liquid steel. According to Körber and Oelsen,[10] carbon primarily influences the activity of iron and manganese by decreasing it, so that the activity coefficient of silicon becomes seemingly higher. The variation of the activity produced by carbon is equal for both iron and manganese. Therefore the value of k_{Mn}^{Fe} remains independent of the carbon content. (The same could be said for the chromium reactions.)

The Equilibria of Oxygen in Acid Open-Hearth Steel

According to Körber and Oelsen,[2] the equilibrium content of oxygen in liquid steel in the pure Fe–Mn–Si–O system depends on the FeO content of the slag. One can use the equation:

$$k_{Fe}^{O} = \frac{[O]}{(FeO)}$$

or, by substituting instead of (FeO) its value from the equations for k_{Mn}^{Fe} or k_{Si}^{Fe}, the expressions:

$$k_{Mn}^{O} = \frac{[Mn][O]}{(MnO)}, \text{ and}$$

$$k_{Si}^{O} = [Si][O]^2, \text{ respectively.}$$

The dependence of k_{Fe}^{O}, k_{Mn}^{O}, and k_{Si}^{O}, respectively, on the carbon content of steel for the temperature range 1590–1640° C. (mean temperature 1625° C.) is shown in Figs. 10 to 12. There are not many determinations of oxygen in liquid acid steel, with full details on the composition of steel and

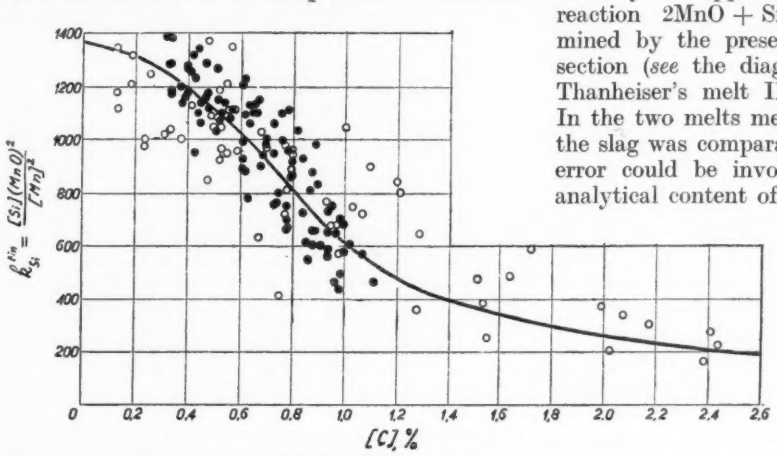

FIG. 9—Dependence of the k_{Si}^{Mn} values on [C]. The solid circles represent the author's data and the clear circles the data of Körber and Oelsen. The drawn curve represents the average values of k_{Si}^{Mn}

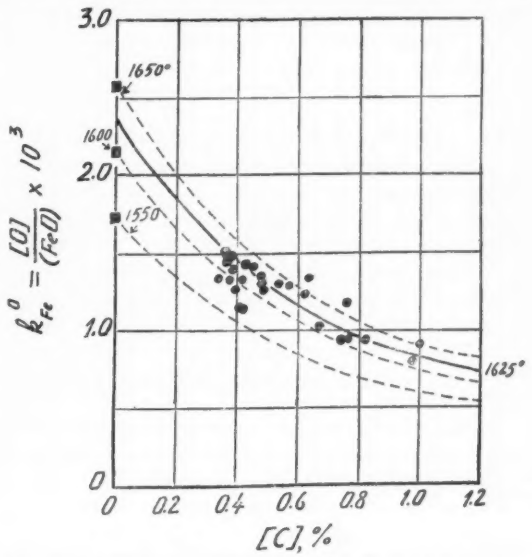

FIG. 10—Dependence of k_{Fe}^{O} on [C]. The points correspond to the mean temperature of 1625° C. The curves for the indicated temperatures were drawn on the assumption that the dependence of k_{Fe}^{O} on temperature in the presence of carbon is the same as in the carbonless Fe–Mn–Si–O system

slag, in the literature. To the best of the author's knowledge, the most reliable data were given by Bardenheuer and Thanheiser.[11] However, the method of steel refining in the heats studied by these authors involved the presence of very viscous slags, so that the bulk of the slag was not in equilibrium with the liquid steel, and the slags in their melts contained excessive contents of ferrous oxide. Only two of their melts (melts II and III in their investigation) can be regarded as fairly well approaching the equilibrium for the reaction $2MnO + Si = 2Mn + SiO_2$, as determined by the present author in the foregoing section (*see* the diagrams for Bardenheuer and Thanheiser's melt III, reproduced in Fig. 13). In the two melts mentioned the FeO content of the slag was comparatively low, so that no large error could be involved in assuming that the analytical content of MnO corresponded practically to its equilibrium value. The data for these two melts are used to supplement the present author's k_{Mn}^{O} values, in Fig. 11.

For the calculation of $k_{Si}^{O} = [Si][O]^2$, Bardenheuer and Thanheiser's data were used, because in the expression for k_{Si}^{O} only the concentrations

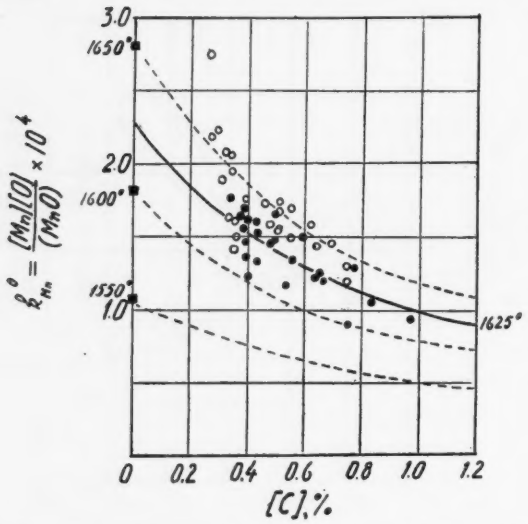

FIG. 11—Dependence of k^{O}_{Mn} on [C]. The solid circles represent the data from Tables II and III (temperature 1625° C.) and the clear circles the data of Bardenheuer and Thanheiser (temperature approximately 1650° C.)

in liquid steel enter (the activity of solid silica is constant). The scattering of the separate values of k^{O}_{Si} is much greater than that for k^{O}_{Mn} or k^{Mn}_{Fe}, because small errors in the oxygen determination are intensified by calculating the second power; besides, the k^{O}_{Si} values are very sensitive to the temperature deviations, as the heat of reaction of $Si + 2O = SiO_2$ is very large. Bardenheuer and Thanheiser did not cite the temperature measurements for separate samples, but merely indicated that the temperature of their melts was in the neighbourhood of 1650° C. The k^{O} values derived from their data lie somewhat above the present author's values, which can be partly attributed to the temperature differences between both sets of values.

In Figs. 10 to 12 the solid black squares at 0% of carbon represent the k^{O} values for the pure Fe–Mn–Si–O system at the temperature indicated, as follows from the work of Körber and Oelsen. The curve of the mean values from the present author's measurements, which corresponds to the mean temperature of 1625° C., was extrapolated to 0% of carbon, and the extrapolated values agree with those valid for the pure system. These diagrams confirm that in the case of oxygen reactions also, carbon exerts a strong influence on the equilibria and on the relative activity of oxygen.

On the assumption that the activity coefficient of silicon does not depend on the carbon content and is equal to unity ($f_{Si} = 1$), it can be calculated that the activity coefficients of manganese and oxygen at 1% of carbon are $f_{Mn} = 0{\cdot}7$ and $f_O = 3{\cdot}4$, respectively. Thus, a 3·4 times smaller concentration of oxygen in liquid steel will suffice to precipitate solid silica at the 1% carbon, than the amount of oxygen needed in carbonless steel.

Utilizing the data used for Figs. 9 to 12, the present author has calculated the curves for silicon and oxygen at 1625° C. at various carbon contents for two cases, *viz.*, (FeO + MnO) = 50 and 40% (*see* Figs. 14 and 15).

A more exact investigation of the influence of carbon on equilibria in liquid steel here described will be of considerable importance for the quantitative understanding of the deoxidation of steel which takes place at the temperatures of solidification. The author is convinced that the refined methods of oxygen determination in liquid steel—coupled with exact determination of silicon and pyrometric measurements of temperature—will furnish the necessary data for computation of the temperature variation of the k^{O}_{Si} values, and so will depict the deoxidation of steel with silicon

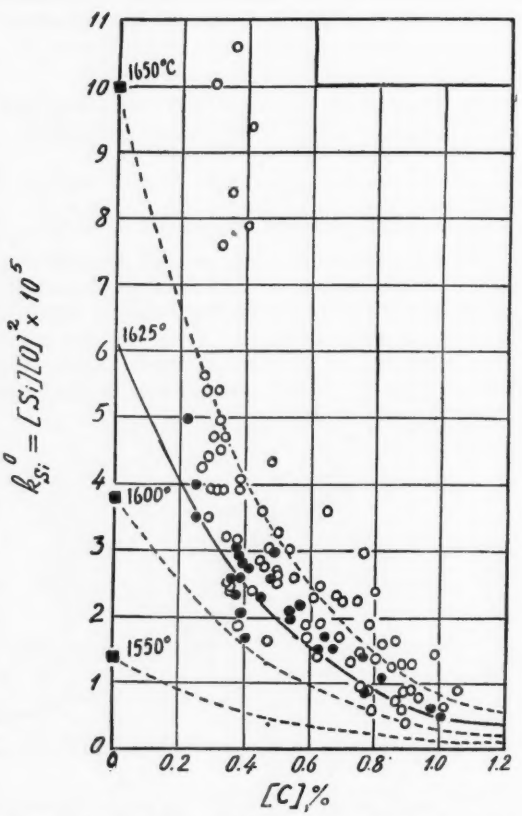

FIG. 12—Dependence of k^{O}_{Si} on [C]. The solid circles represent the author's data (mean temperature 1625° C.) and the clear circles the data of Bardenheuer and Thanheiser (temperature approximately 1650° C.)

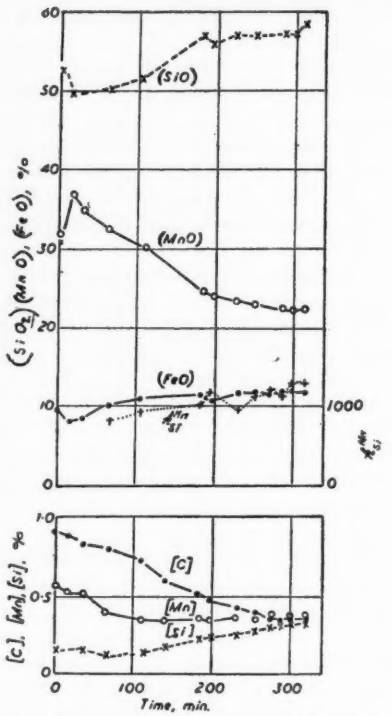

FIG. 13—Data for Bardenheuer and Thanheiser's melt III

and manganese and the nature of deoxidation products at various contents of carbon. It may also be expected that other deoxidation reactions (*e.g.*, with aluminium) will also be strongly influenced by carbon.

In concluding this section of the paper, the author can say that there are two outstanding factors which are operative in producing the deviation of industrial melts from the behaviour of the pure Fe–Mn–Si–O system. These factors are : (1) The influence of carbon on the activity of elements dissolved in steel, and (2) the displacement of equilibria by the admixtures to slag (the effect of dilution).

The Mechanism of Carbon Oxidation

Most of the data on the course of acid open-hearth heats indicate that the reaction C + O → CO is always considerably distant from the equilibrium of this reaction as determined by Nacher. The product [C] [O] is always greater than the equilibrium constant.

Körber and Oelsen[1] assumed that the primary reaction between carbon and oxygen proceeds very rapidly with the formation of the supersaturated solution of carbon monoxide. According to these authors the slowest process in the carbon reaction is the evaporation of the carbon monoxide from such a supersaturated solution in steel.

The reaction can be described in a different manner by assuming that the slowest reaction is the primary reaction between carbon and oxygen, which requires a considerable activation energy, similar to the case of the formation of hydrogen molecules from the electrodeposited hydrogen atoms in the well-known phenomenon of overvoltage.

Körber and Oelsen, in assuming an instantaneous reaction between carbon and oxygen, have come to the conclusion that the oxygen content of open-hearth melts is governed exclusively by the carbon reaction and that there is therefore no equilibrium between the oxygen in steel and the iron oxide in the slag. The present author is convinced that the data so far given are quite conclusive in proving that there is an equilibrium distribution of oxygen between steel and slag, which is strongly influenced by the presence of carbon, as in the silicon reactions.

The present author therefore accepts the second

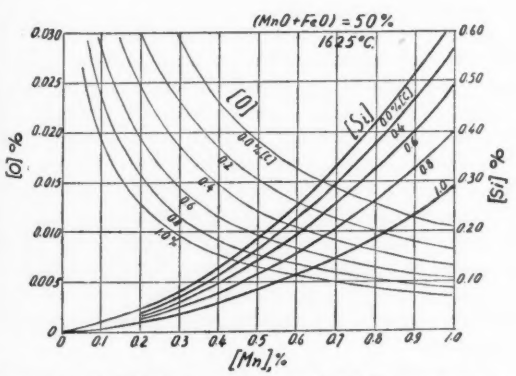

FIG. 14—Equilibrium oxygen and silicon contents at various carbon and manganese contents in steel which is in equilibrium with slags containing 50% (FeO + MnO). Temperature 1625° C.

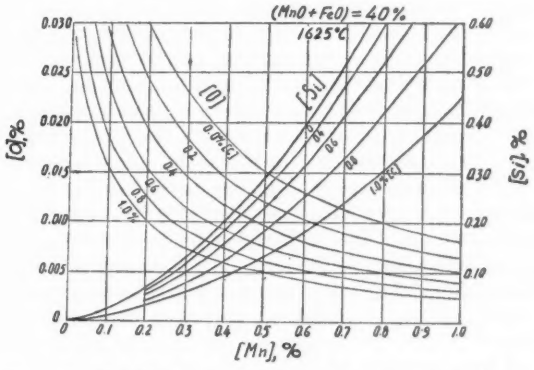

FIG. 15—Equilibrium oxygen and silicon contents at various carbon and manganese contents in steel which is in equilibrium with slags containing 40% (FeO + MnO). Temperature 1625° C.

assumption, *viz.*, that the primary reaction between carbon and oxygen is the slowest process requiring an activation energy. As a consequence, the oxygen content of steel has no connection with the carbon reaction, being influenced only indirectly in its activity by the dissolved carbon.

The activation energy necessary for the oxidation of carbon can decrease by adsorption on crystalline surfaces, such as those of solid silica or solid iron crystals. Taking into account all the observations mentioned in this paper, it is possible to suggest that the course of the reactions in the acid open-hearth furnaces may be as follows. Soon after the melting-out of the charge, an equilibrium is established between the bath and slag with respect to iron, manganese, silicon, and oxygen (an apparent lack of equilibria might in some cases be due to the presence of a high content of silicon in the steel, so that at the slag–metal interface a layer of highly viscous slag with a high content of silica forms). The equilibria are attained especially easily with melts containing higher contents of manganese (from approximately 0·35% and over). Owing to the presence of high contents of carbon at the beginning of liquid-steel refinement, the equilibria are displaced in comparison with the pure (carbonless) Fe–Mn–Si–O system, as already shown. The start of the carbon oxidation depends mainly on the properties of the furnace hearth under the metal bath. As the whole surface of the hearth under the metal is covered with a film of fluid slag (and this occurs very readily at high contents of manganese), the reaction of carbon is strongly hampered. The course of a heat investigated by Malcor (*see* Fig. 6) may serve as an example. In this heat, the manganese content of which was high, more than 5 hr. elapsed before the oxidation of carbon began to proceed at a measurable rate. After the commencement of carbon oxidation, the mechanical action of the boil makes direct contact of the liquid steel with the solid silica of the hearth easier, so that the reaction is accelerated. At the same time, the mixing action of the boil aids in equal distribution of the constituents in both liquid phases and in the attainment of equilibria (of course, when an excessive amount of viscous slag is present from the start, the establishment of equilibria can be localized only to the thin interface layer). The boil also aids in raising the temperature of the bath and in bringing the slag to saturation with solid silica. All these factors (the lowering of the carbon content, the rise of temperature, and the change in the slag composition) displace the positions of equilibria for reactions of iron, manganese, silicon, and oxygen.

The rôle played by physical conditions can be substantiated by the following observations. The author determined the mean rate of carbon oxidation during a furnace campaign in which the temperature regime of the furnace was practically the same for all heats. The slag contained only small amounts of foreign oxides (about 5% of $CaO + Al_2O_3$), and no additions to the slag were made during the refining. The time required to eliminate 0·4% of carbon (from 0·9 to 0·5%) was determined for more than 50 heats. The manganese content of the bath usually became constant in the middle of the boiling period. It was found that the heats with 0·45–0·55% of manganese boiled, on the average, more slowly than those with higher and lower contents of manganese. The time required for the elimination of 0·4% of carbon at 0·5% of manganese was about $3\frac{1}{2}$ hr., which was almost twice as great as that at 0·35% of carbon and 0·65% of manganese. At manganese contents lower than 0·4%, the rate of carbon oxidation was approximately proportional to the content of iron oxide in the slag.

The phenomena observed could be explained as follows : A film of fluid slag always forms on the surface of the solid silica hearth under the steel bath. The fluidity of this film probably increases with increasing content of manganese oxide (and therefore with the manganese content of the steel with which the slag film is in equilibrium). The optimum conditions at which this film forms a continuous layer on the surface of solid silica correspond to 0·5% of manganese in the steel. With lower manganese contents the slag is too viscous to form a continuous film on the surface of solid silica, and at higher manganese contents the slag layer is too fluid and has a high surface tension, so that it easily collects into drops. In both cases the surface of the solid (crystalline) silica is more accessible for the metal than when the film is in equilibrium with 0·5% of manganese in the steel. Therefore the carbon oxidation is more catalysed at both low and high contents of manganese.

At low contents of manganese (below 0·35%) the rate of carbon oxidation may become larger than the rate of oxygen transfer from the slag. Carbon then reacts directly with the oxygen of the solid silica, thus : $2C + SiO_2 \rightarrow Si + 2CO$, and therefore the silicon content in the steel increases more rapidly than the establishment of equilibria between the slag and the bulk of the steel. Silicon is readily oxidized on reaching the upper layers of the steel in contact with the slag, forming a highly viscous layer of slag, which further prevents the participation of the bulk of slag in reactions with liquid steel.

However, if the slag is kept fluid (*e.g.*, by additions of limestone), then it retains its ability to determine the course of all the reactions (excepting the oxidation of carbon).

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HISTORICAL NOTE No. 5

The Black Death of 1348–49 and Its Effect upon the Iron Industry*

By Dr. H. R. Schubert†

An event of the most far-reaching effect upon public and economic life in the Middle Ages was the great pestilence commonly known as the Black Death, a most virulent form of bubonic plague. Originating in the East and travelling across Europe, the Black Death first entered England in August, 1348, through southern ports, and rapidly swept over the entire country. Although the virulence of the plague had abated by the late autumn of 1349, the pestilence remained, periodically reappearing as a sequel to famine caused by bad harvests ; its last occurrence was the Great Plague of London, in 1665.[1]

The most striking features of the first visitation were the suddenness of the outbreak and the appalling mortality. About 1½ millions died out of a total population of about 4½ millions, and the plague became the greatest catastrophe in the history of England. The decrease in man-power which quickly resulted from the terrific death-roll created a shortage of labour such as had never before been experienced. Production fell accordingly, and the decline continued steadily in the years immediately following the first wave of the pestilence. For instance, at the ironworks at Tindale, in Cumberland, the output of blooms for the year 1333 had been 294 ; the outputs for the years 1350, 1351, and 1353 were 204, 143, and 26 blooms, respectively.[2]

Another result of the shortage of man-power was a general rise in prices, noticeable in agriculture as well as in the crafts and in industry. Particulars as to the effects of this on the iron industry are provided by the accounts for the Tudeley ironworks,[3] for the periods 1329–34 (before the plague) and 1350–54 (after the plague). The rise in prices as between the two periods begins with the cost of production. In the first period the carriage of iron ore dug in the adjacent forest or chase of Southfrith (between Tonbridge and Tunbridge Wells) cost about 18*s.* per 100 blooms produced, and the cost of bringing charcoal to the works was from 3*s.* 6*d.* to 4*s.* In the second period these costs had increased to 27*s.* for the ore and to between 8*s.* 7*d.* and 9*s.* 2*d.* for the charcoal. The average price of iron rose accordingly. Previous to the plague this had been fairly constant at 1*s.* 8*d.* per bloom, but afterwards the price rose to, in some cases, more than double this figure. The ironworkers' piecework rate of 5½*d.* per bloom increased to between 7½*d.* and

* Received 8th August, 1947.

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[1] G. M. Trevelyan, "History of England" (second edition), p. 237. London, 1937.

[2] T. Rogers, "A History of Agriculture and Prices in England," vol. 1, p. 472. Oxford, 1866.

[3] The Tudeley ironworks were situated east of Tonbridge, Kent. Details of the accounts have been given by M. S. Guiseppi, *Archaeologia*, 1913, vol. 64, pp. 145–164.

9½*d.* over the same periods. The average prices and wages thus approximately doubled.

The Tudeley ironworks were leased to Richard Colepeper some time after 1354, but they appear to have been abandoned following the second wave of the plague, in 1360–61. There are indications that two other ironworks—leased to one Thomas Hardy—in the neighbourhood had already been abandoned after the first outbreak.

When the great consumption of iron by the mainly agricultural population of the realm at the time is considered, the sudden rise in price was bound to have a serious economic effect. The government, well aware of the gravity of the crisis, attempted to alleviate the situation by legislation aiming at control of labour and prices, and at prohibition of the export of iron. By a statute of 1354 the export of iron was forbidden and its sale at excessive prices rendered punishable. Chapter V of the statute enacted that "iron made in England and iron brought into England and there sold shall not be carried out of the said realm of England upon pain of forfeiting the double to the King," and that "Justices whom the King will thereto assign, shall have power to enquire of them that sell the iron at too dear a price, and to punish them after the quantity of the trespass."[4]

The statute illustrates the gravity of the crisis, but it no more succeeded in controlling prices than did other enactments succeed in controlling labour. All over the country the labourers realized the opportunities offered to them by the shortage of man-power, and crowds left their places of employment in search of higher wages. This caused the first great struggle in English history between the rich and the poor, between capital and labour. But in this struggle the government mainly took the part of capital, although it did try to keep the price of food, and in particular of bread, meat, and fish at a low level.[5]

A Royal Proclamation of 1349 ordered all labourers to remain at their places of employment till their contracts expired, and to work at the old rate of wages. As the terms of this Proclamation were not sufficiently observed they were embodied in an Act of Parliament in 1351, the so-called Statute of Labourers. This enforced upon labourers the acceptance of work when it was offered at the rate of wages customary before the coming of the Black Death. It imposed heavy fines on those who disobeyed, and imprisonment on workers leaving their places of employment. On the other hand, by means of the same statute, an attempt was made to restrain the rise in food prices by the prohibition of sale for "excessive" gain.

The Statute of 1351 is the first experiment in controlling labour by the government. As it was not successful in achieving its purpose, the statute had to be re-enacted again and again, and the already heavy penalties were made more severe. By the re-enactment of 1360, labourers who broke their engagement and left their places of employment in search of higher wages were outlawed; when caught they were to be branded on their foreheads with the letter **F**, as a token of their "falsity." Towns which sheltered them were to be fined. However, despite all this legislation, the dearth of labour continued, especially in the mines, and in 1373–74 miners and coal-bearers had to be forcibly seized for work in the coal mines of Wickham and Gateshead, on the southern bank of the Tyne, in County Durham.[6]

It is a remarkable phenomenon that steel was considerably less affected by the general rise in prices. Compared with iron, which increased in price by more than 100% immediately after the Black Death and remained at that high level for the second half of the century, steel was only 25% dearer than before. For a better understanding of this disparity it is necessary to take into consideration that in the first half of the fourteenth century, *i.e.*, before the coming of the Black Death, the market price of steel was about four times as high as that of iron.[7] Steel was mainly an article of import from abroad, brought into the country by the merchants of the Hanseatic League.[8] This expensive material was employed for the manufacture of weapons, armour, and finer tools and instruments. The commoner types of tools used in agriculture and by the craftsmen in the towns could be manufactured from iron produced locally and given a character akin to steel by repeated forgings in the smithies, thus freeing it from impurities. The price of steel, already high before the first outbreak of the pestilence, thus did not soar nearly as much as the price of iron, which was in much greater demand. It may even be inferred from the disparity in the prices that the demand for the finer varieties of implements was considerably smaller as a result of the decline in civilization in the economic crisis caused by the Black Death, when the stringencies necessitated restriction in demand and produce to the sheer necessities of life.

[4] "The Statutes of the Realm," vol. I, p. 345. London, 1810.

[5] G. T. Warner, "Landmarks in English Industrial History," chapter VI, "The Black Death," p. 81 *et seq.* Glasgow, 1926.

[6] Victoria County History of Durham, vol. II, p. 322. London, 1907.

[7] T. Rogers, "A History of Agriculture and Prices in England," vol. I, p. 474.

[8] For information as to the importation of steel in the fourteenth century, *see* L. F. Salzman, "English Trade in the Middle Ages," p. 363. Oxford, 1931.

The Mechanical Properties of a Nickel–Chromium–Molybdenum Steel Obtained by Stepped Quenching*

By G. Delbart† and R. Potaszkin‡

SYNOPSIS

The purpose of this investigation was to compare the mechanical properties of a nickel–chromium–molybdenum steel after stepped quenching, with those obtained after air cooling or oil quenching followed by tempering.

First of all, the **S** *curve of this steel was determined. Its comparison with that of a chromium–molybdenum steel studied previously shows that the addition of nickel is effective in depressing the temperatures at points of inflexion of the* **S** *curve by an amount practically equal to that of the depression of the transformation points and, in particular, in increasing the stability of austenite.*

The influence of the temperature of the intermediate quenching bath and of the tempering temperature on the various mechanical properties was studied, and these were then plotted against the tensile strength. The best results were obtained with martensitic quenched structures, which can be obtained by direct oil quenching or air cooling, and also by stepped quenching with intermediate holding at 200° C. or between 400° and 600° C. As far as the range of tensile strength between 90 and 110 kg./sq. mm. is concerned, similar results can also be obtained by holding in the bainitic region (about 300°–350° C.) and tempering. Test-pieces of bainitic structure tempered at low temperatures and possessing a tensile strength of over 140 kg./sq. mm. exhibit a higher impact strength and a lower elastic limit than those with a martensitic quenched structure and the same tensile strength.

The structure resulting from stepped quenching with holding for 6 hr. in the pearlitic range (about 650° C.) comprises two different components (troostite and martensite), and all mechanical properties are considerably inferior to those obtained with any of the other treatments.

In a previous communication[1] we reported the practical results obtained on a chromium–molybdenum structural steel after stepped quenching. The essential aim of this work was the study of the variation of the mechanical properties with the different conditions under which this quenching was carried out, and in a general way, the investigation of a range of applications of heat treatments. In the present work, we communicate the results of a similar investigation carried out on another structural steel, of the nickel–chromium–molybdenum type.

Since our earlier paper appeared, H. Jolivet and R. Castro[2] have made a new and important contribution to the general knowledge of the principal factors governing the isothermal decomposition of austenite. Elsewhere,[3] the recent American bibliography contains numerous references on the subject of the establishment of **S** curves for steels of different compositions constituting an experimental investigation of the greatest importance.

Since the work of Bain, and until recently, the great majority of papers on this subject have been devoted to new information about the general phenomenum of the decomposition of austenite. On the other hand, there is today, only a small amount of accurate information emphasizing the actual interest of the practical application of stepped quenching to structural steel. It appears that there exist definite possibilities, as is shown by the use in the U.S.A. of stepped quenching under the names of austempering and martempering.

In our first paper, which dealt with a steel containing 0·30% of carbon, 2% of chromium, and 0·5% of molybdenum, we demonstrated the existence in the bainitic region, of the two characteristic temperatures of isothermal quenching, 500° and 400° C. The first of these allows us to obtain the higher mechanical properties, and the second corresponds to the lower mechanical properties. Structures formed below 300° C. are

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martensitic in type and have properties not greatly different from those obtained by direct oil quenching. The decomposition of austenite at 650° and above, gives rise to softer constituents which correspond to the annealed condition (pearlitic region).

As regards the comparative effects of different treatments, we have established that by stepped quenching at 500°–525° C., with final air cooling and without tempering, a high tensile strength (150 kg./sq. mm., approx.) is obtained, together with an impact strength equal or superior to that obtained by a classical treatment (oil quenching and tempering) with the same tensile strength. On the other hand, the highest mechanical properties, corresponding to a tensile strength of from 130 and 105 kg./sq. mm., are obtained by oil quenching and tempering. Finally, a quench stepped at the proper temperature, followed by tempering gives, for a tensile strength of 105 and 75 kg./sq. mm., a greater impact strength and as good an elastic limit as those resulting from an oil quench followed by tempering.

Stepped quenching then, apart from the greater safety which it offers for the heat-treatment of samples liable to quenching cracking and deformation, and residual stresses, is a method which permits the realization of mechanical properties comparable or superior to those obtained by oil quenching, when one wants to obtain a tensile strength greater than 130 kg./sq. mm., or less than 105 kg./sq. mm.

Similar conclusions may be drawn from the present paper, of which the only purpose is to communicate some new experimental facts capable of practical application.

I—Conditions of Test

The steel chosen for this study had the following chemical composition :

C, %	Si, %	P, %	S, %	Mn, %
0·35	0·25	0·021	0·010	0·43

Ni, %	Cr, %	Mo, %
2·52	1·55	0·53

The Ac_1 transformation point is at 720° C., and the Ac_3 transformation takes place at 770° C. This steel was elaborated by Anciens Etablissements, Cail, at Denain (Nord), in an acid open-hearth furnace of 25 tons capacity.

The austenite grain-size at a quenching temperature maintained at 850° for 30 min. is No. 5 A.S.T.M. The austenite grain-size under the conditions of the MacQuaid-Ehn test is No. 3 A.S.T.M.

The different test samples used were bars of 25 mm. dia. obtained from a 10-ton ingot. All the bars were given a preliminary normalizing heat-treatment consisting of annealing at 850° for 30 min., and cooling in air, followed by reheating to 650° for 2 hr., with furnace cooling.

Determination of S *Curve*

As in the previous paper, the isothermal decomposition of the austenite was followed by microscopic examination of small samples which had been subjected to reheating to 850° followed by quenching in a salt or lead bath, maintained at a constant temperature for different times and different temperatures, and water quenched.

The temperatures used for the intermediate bath varied from 200° to 700° C. in 50° steps. The length of time of immersion at that temperature varied from 1 min. to 60 hr., and even reached 170 hr. (about 1 week) for some specimens.

Mechanical Tests

After the normalizing treatment described above, the bars were cut into test specimens with an overall thickness of 0·3 mm. ; the latter were treated in batches, each comprising a tensile test-piece of 13·8 mm. and three samples for Mesnager tests, in the conditions detailed below. The test samples were brought to their final condition by rectification after treatment.

The quenching conditions were as follows :

(*a*) Heating at 850° C. for 30 min.—Cooling in calm air,
(*b*) „ „ „ „ „ „ Cooling in oil,
(*c*) „ „ „ „ „ „ Cooling in intermediate bath, with 6 or 1 hr. isothermal treatment at temperatures between 200° and 700° C. in 50° steps, followed by cooling in calm air.

The tempering conditions were :

(*a*) Stress-relieving heat-treatment for 6 hr. at 200° C.
(*b*) Tempering for 2 hr. at 500° C.
(*c*) Tempering for 2 hr. at 600° C.
(*d*) Tempering for 2 hr. at 700° C.

In addition, a batch of samples was subjected to 30 min. annealing at 850° followed by slow furnace cooling without tempering. In all, therefore, there were 73 batches of specimens, each batch having undergone a different heat-treatment.

II—Study of the Decomposition of Austenite

S *Curve*

In Fig. 1, which represents the decomposition curve of austenite determined as indicated above, five characteristic temperature ranges can be distinguished :

(i) Above 650° C. Deposit of free ferrite at the grain boundaries of the austenite.
(ii) 650–600° C. ... Pearlite region.
(iii) 550–400° C. ... Region of stable austenite.
(iv) 400–250° C. ... Intermediate region.
(v) Below 250° ... Martensite region.

Figures 2 to 10 show the photomicrographs obtained under a linear magnification of 1250,

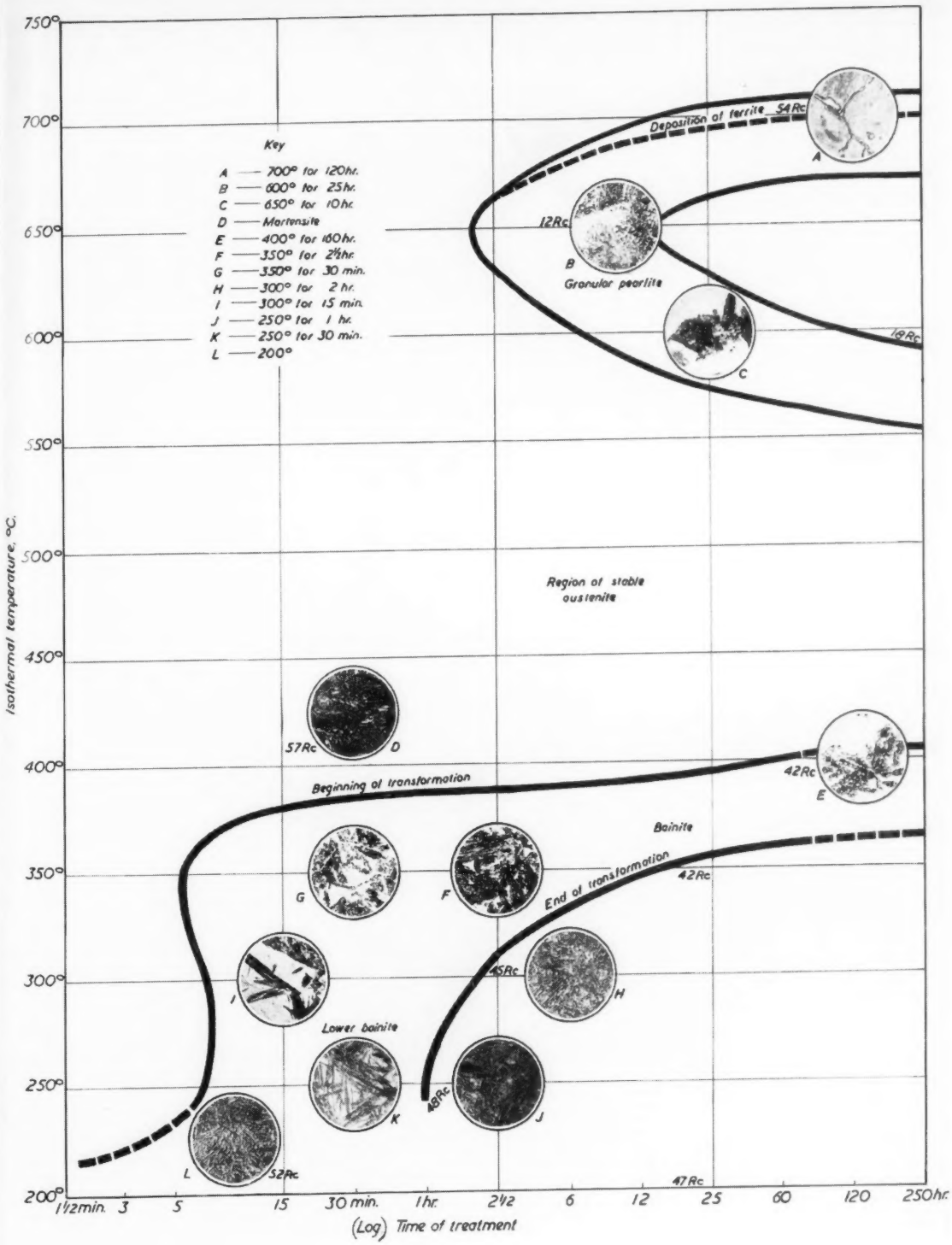


FIG. 1—Curve representing the isothermal transformation of austenite in a nickel-chromium-molybdenum steel quenched from 850° C. and possessing grain-size No. 5

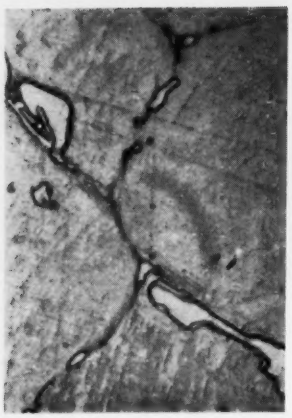


FIG. 2—Ferrite deposit, 700° C.

FIG. 3—Troostite, 650° C.

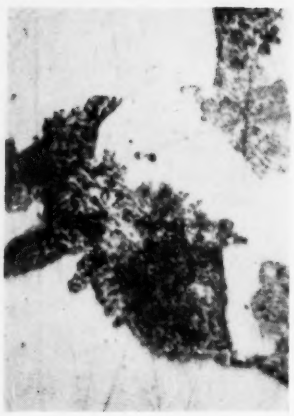


FIG. 4—Granular martensite, 600° C.

FIG. 5—Upper bainite, 400° C.

FIG. 6—Bainite, 350° C.

FIG. 7—Lower bainite, 300° C.

FIG. 8—Lower bainite, 250° C.

FIG. 9—Structure formed at about 200° C.

FIG. 10—Tetragonal martensite.

FIGS. 2–10—Types of structure formed on the isothermal decomposition of austenite in a steel containing 0·30% of carbon, 2·5% of nickel, 1·5% of chromium and 0·5% of molybdenum. Structures preserved by water quenching after treatments. Etched with alcoholic nitric acid × 1250

of some of the characteristic structures formed at different temperatures. The austenite which was not decomposed during the isothermal treatment was transformed into martensite during water quenching. The conditions of formation of these structures call for further detailed information:

The precipitation of free ferrite which takes place in the region of 700° C. is not accompanied by the formation of any carbide, even after a week at that temperature (Fig. 2).

The transformation of austenite in the pearlitic range only begins after a long period of incubation (more than 2 hr.), giving a two-phase constituent. At 650° the latter presents locally, a lamellar appearance which is not very pronounced (troostite, *see* Fig. 3); at 600° C. its structure is uniformly granular (granular pearlite, *see* Fig. 4).

There is no evidence of any change during a week's isothermal treatment at 500°.

The acicular character of the structures of the intermediate region becomes more marked the lower the temperature of formation. These constituents are, namely, upper bainite towards 400° C. (Fig. 5), bainite at 350° (Fig. 6), and lower bainite between 300° and 250° (Figs. 7 and 8).

The constituent formed by stepped quenching at 200° appears to be martensitic in nature (Fig. 9). As we have already found for a chromium–molybdenum steel, the hardness of the martensite decreases appreciably with increase in the time of holding at 200° (57 Rockwell 'C' (Rc) after a few seconds, 52 Rockwell 'C' after a few minutes, and 47 Rockwell 'C' after 25 hr), although no structural modification is visible under a magnification of 1250.

Influence of Nickel on the Decomposition of Austenite

The isothermal transformation diagram of this steel differs markedly from that of the molybdenum–chromium steel investigated in our previous communication, *see* Fig. 11. It appears that these differences should be attributed to a large extent to the presence of nickel, which was the principal difference between the two steels. Their chemical compositions were:

	C, %	Si, %	Mn, %	Ni, %	Cr, %	Mo, %
Steel *A* ...	0·31	0·26	0·46	0·26	2·08	0·48
Steel *B* ...	0·35	0·25	0·43	2·52	1·55	0·53

The essential differences between the **S** curves for these two steels may be summarized as follows:

The three inflexions of the **S** curve for steel *A*—the Ar_1 bend in the pearlitic region, the bend corresponding to the greatest stability of austenite, and the Ar_2 bend of the intermediate region—are not found at the same temperatures in steel *A* as in steel *B*. Indeed,

TABLE I—*Characteristic Temperatures of Steels* A *and* B

Arrest	Steel *A*	Steel *B*	Difference
Ac_1	790°	720°	70°
Ac_3	835°	770°	65°
Ar_1	720°	650°	70°
Stability of austenite	about 560°	about 475°	about 85°
Ar_2	425°	350°	75°

the characteristic temperatures of the three inflexions in steel *B* are lower than those in steel *A* by an amount approximately corresponding to the difference between their respective allotropic transformation points, as is shown in Table I.

The incubation period is substantially longer for nickel steel *B* than for steel *A*. The influence of the greater stability of the austenite, which is already known, on the behaviour of nickel steels subjected to heat-treatments, will at once be seen.

As a result of the lowering of the transformation temperature, the pearlitic region of steel *B* does not include the lamellar constituent of the classical pearlitic type. On the other hand, the extension of the range of stability of the austenite reduces the extent of the intermediate region. Upper bainite can only be formed, therefore, in a narrow range of temperatures (about 400°) and after more than 25 hr. incubation.

The effect of nickel on the form of the **S** curve explains the influence of this element on the suitability for quenching, and the manner in which it decreases the critical cooling rate.

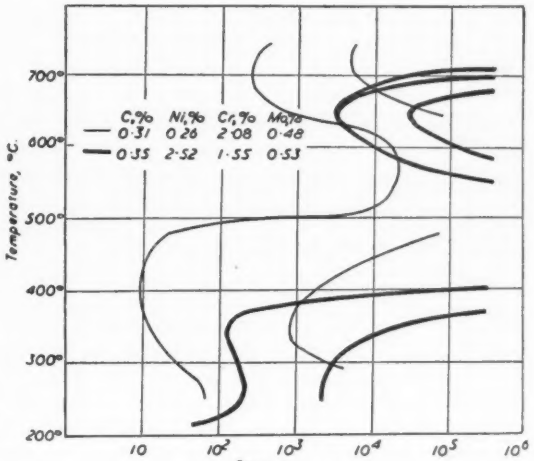


FIG. 11—Comparison of the **S** curves of a Ni–Cr–Mo steel and a Cr–Mo steel

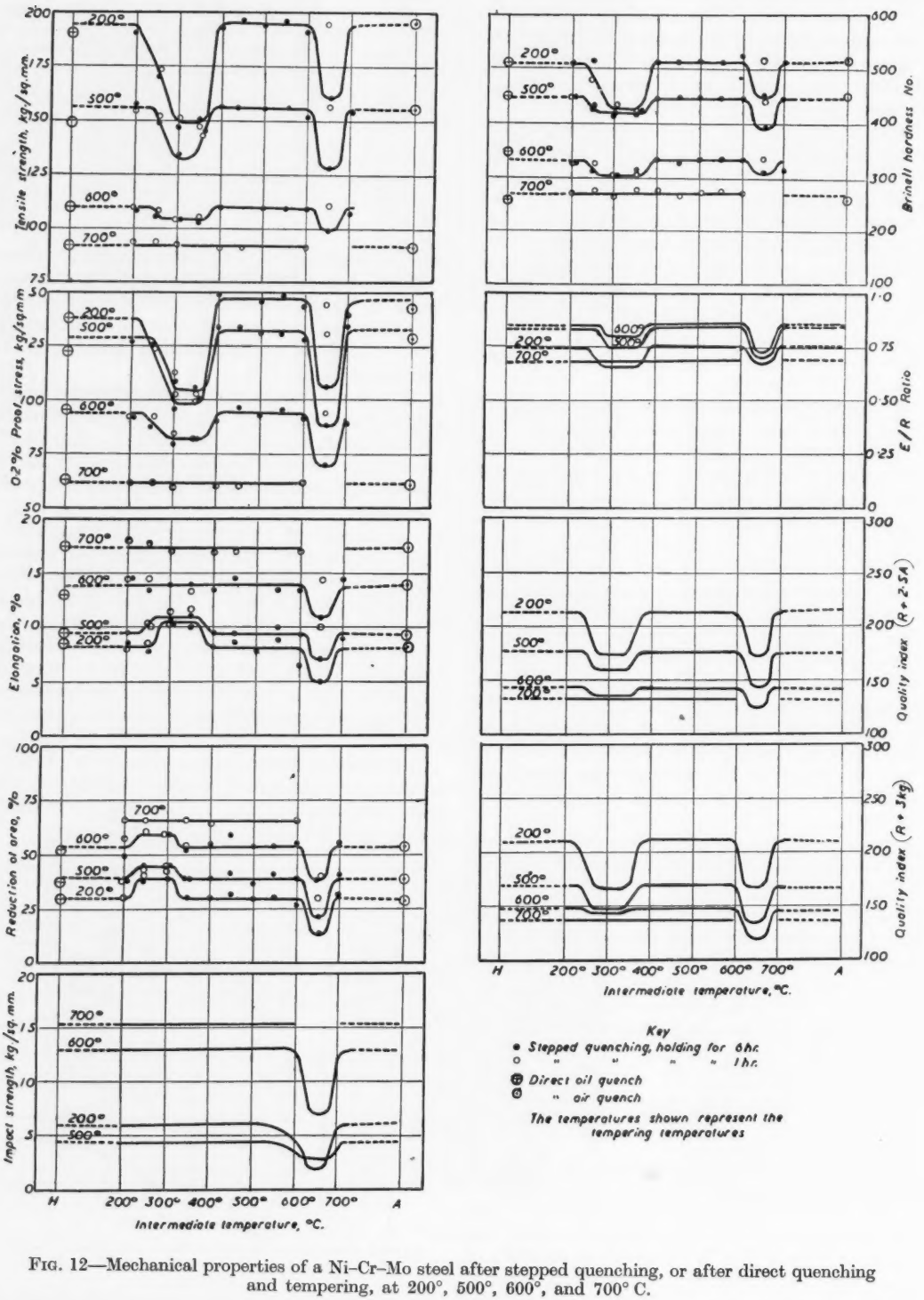


FIG. 12—Mechanical properties of a Ni–Cr–Mo steel after stepped quenching, or after direct quenching and tempering, at 200°, 500°, 600°, and 700° C.

The transformation of austenite in the case of the nickel–chromium–molybdenum steel can only commence in the pearlitic region after at least 2 hr. heating (at about 650°), and for the formation of a fine pearlitic structure of Brinell hardness less than 200 the temperature requires to be held isothermally for 12 hr. Austenite in the undecomposed pearlitic region can give rise to bainitic constituents below 400°, but in that region the transformation is likewise fairly slow, and decomposition takes over an hour to complete. Also, during continuous cooling under the usual quenching conditions, the austenite will decompose either to martensite with a hardness of 54 to 57 Rockwell 'C,' or to a bainitic constituent of hardness greater than 42 Rockwell C. The exact nature of the structure formed will depend on the cooling rate, the dimensions of the specimen, and the part of the specimen considered.

III—Mechanical Tests

The results of mechanical tests on specimens subjected to the different heat-treatments described above, are collected on the graphs of Fig. 12. They are plotted as functions of the temperature of the intermediate bath, and are for a 6-hr. isothermal heat-treatment with air cooling. Each mechanical property is represented by a family of 4 curves, each of which corresponds to a different tempering temperature, *viz.*, 200°, 500°, 600°, and 700° C. On the same graphs are also shown the points corresponding to 1 hr. isothermal treatment, together with those obtained by direct oil quench or air cooling.

Before entering into a discussion of each of the properties investigated, it is convenient to refer to a few interesting properties of a general nature shown by the results. They are, it is clear, logically connected with the conditions of isothermal decomposition of austenite discussed above.

Stepped quenching with maintained isothermal heat-treatment at temperatures equal to or higher than 400° C. gives, except at the temperature of 650° C., mechanical properties similar to those obtained by oil quenching or air cooling. The few discrepancies observed in one direction or the other are of the order of magnitude of the spread of the experimental errors. In this temperature region and for periods up to 6 hr., austenite undergoes no decomposition, but changes into martensite during the course of the subsequent cooling, whatever mode of quenching is used. This statement applies just as well to samples of martensitic structure tempered at 200°, as to those of sorbitic structure which have been tempered at a higher temperature.

The mechanical properties obtained after stepped quenching with isothermal holding for 6 hr. at 650°, are appreciably inferior to those obtained by direct quenching. This is explained by the presence of troostite (indicating incomplete quenching) formed by the decomposition of austenite, which starts after about 2 hr. at this temperature. This inferiority after quenching from 650° remains after tempering. On the other hand, the mechanical properties resulting from 1 hr. at this temperature are quite similar to those obtained by direct quenching, which is in agreement with the shape of the **S** curve.

The mechanical properties resulting from a stepped quenching with isothermal heat-treatment in the intermediate region of 250–350° are, in general, different from those obtained by direct quenching, but the differences decrease as the tempering temperature is increased. In this temperature range the decomposition of austenite is relatively rapid, and the structures obtained by stepped quenching include a certain proportion of bainite. After reheating to 700°, the mechanical properties are practically independent of the method of quenching, and microscopic examination shows that the same structures are obtained.

It is also found that increasing the length of the time of isothermal heat-treatment from 1 to 6 hr. in the intermediate region, has no appreciable effect on the mechanical properties. It is known that the decomposition of the austenite proceeds most rapidly in the initial stages, and then continues much more slowly.

Tensile Strength and Hardness

The curves showing the variation of tensile strength and hardness as functions of the temperature of the intermediate bath are particularly representative. They consist of segments of horizontal straight lines corresponding to an initial martensitic state, with two dips; one of these is between 250° and 400° C., and corresponds to an initial bainitic structure, and the other, between 600° and 700° C., corresponds to an initial structure of martensite and troostite.

The bainitic constituents formed at 350° and 300° C. are less resistant and softer than those formed by isothermal heat-treatment at 250° C.

As stated above, the differences diminish as the tempering temperature increases, and disappear almost entirely when this temperature reaches 700° C., a temperature near but below the Ac_1 point.

XUM

Elastic Limit and E/E *Ratio**

The variations of elastic limit with temperature of intermediate bath are, on the whole, similar to those of the tensile strength. On investigating the effect of temperature of tempering, an overlapping of certain curves is found which is more easily examined by considering the variations of the E/R ratio.

The value of this ratio undergoes changes similar to those of the properties discussed above, and in the same temperature regions. There is thus a decrease in E/R in the intermediate region and in the region of incomplete quenching. The influence of tempering temperature is more complex ; the maximum values are obtained with a tempering temperature of 600°, and the minima according to conditions, with a tempering temperature of 700° or 200°. This particular behaviour is explained for different constituents by the different trends of the curves in which E/R is plotted against the tempering temperature. The decrease in E/R after tempering at a temperature near the Ac_1 point is observed generally in all steels.

Elongation at Fracture, and Quality Index $R + 2{\cdot}5A$*

For a given tempering temperature the elongation is almost independent of the temperature of the quenching bath, provided that quenching takes place in the martensitic region. Some more or less important fluctuations, due partly to variation in strength, are seen in both directions in the intermediate and pearlitic regions. These results may, however, be more easily interpreted by considering the curves of quality index.

An appreciable reduction of quality index is found in the two non-martensitic regions of quenching. The differences diminish, however, after tempering at increasing temperatures, and vanish almost completely after tempering at 700°.

Reduction of Area

The general trend of the reduction-of-area curves is in all respects similar to that of the elongation curves. The values corresponding to quality index $R(100/100 - \sigma)$ show a fairly large scatter and could not easily be plotted ; they are not, therefore, shown in Fig. 12.

Impact Strength and Quality Index $R + 3K_M$*

For a given tempering temperature, the variation of impact strength with temperature of the intermediate bath is small, provided that the temperature is in the martensitic or the intermediate region, *i.e.*, not in the region in the neighbourhood of 650°. In this region the differences between values are of the same order of magnitude as the spread of results.

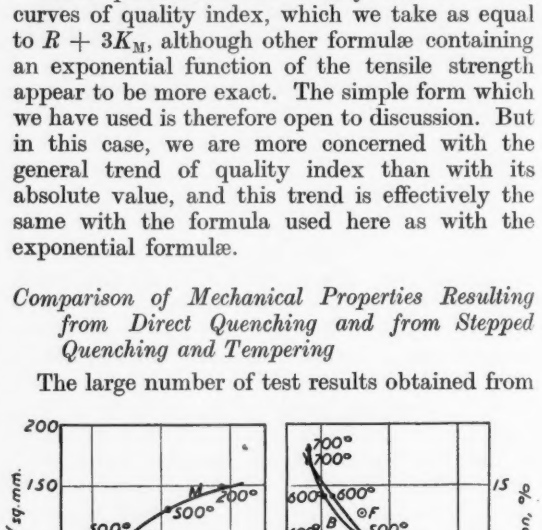
It is to be noted that, despite their apparent similarity, the results are not comparable one with another ; for the impact strength being equal, the tensile strength of the samples consisting of martensite is, for low tempering temperatures, considerably greater than that of samples of bainitic structure.

For isothermal heat-treatment at 650° (incomplete quenching) the values of impact strength are considerably reduced, despite the corresponding reduction in the tensile strength.

These phenomena are clearly shown in the curves of quality index, which we take as equal to $R + 3K_M$, although other formulæ containing an exponential function of the tensile strength appear to be more exact. The simple form which we have used is therefore open to discussion. But in this case, we are more concerned with the general trend of quality index than with its absolute value, and this trend is effectively the same with the formula used here as with the exponential formulæ.

Comparison of Mechanical Properties Resulting from Direct Quenching and from Stepped Quenching and Tempering

The large number of test results obtained from



F = Furnace cooled B = Bainitic condition
M = Martensitic condition H = Troostitic condition

FIG. 13—Comparison of the mechanical properties of a Ni–Cr–Mo steel in the martensitic, bainitic, and troostitic conditions

* E = Yield point, R = tensile strength, A = elongation, and K_M = impact strength (measured with a Mesnager notched test-piece).

the various heat-treatments make their comparison difficult. What is of practical importance, is to be able to compare the tensile strength with the values of the other mechanical properties such as elastic limit, impact strength, elongation, and reduction of area, obtained by different heat-treatments. We have attempted to do this by collecting, in Fig. 13, the results of tests representative of the general behaviour, shown as a function of the tensile strength.

To facilitate interpretation and to simplify their representation on a graph, the results have been re-arranged as described below.

Considering the graphs of Fig. 12, it has been assumed that the mean mechanical properties were the same for all temperatures of the intermediate quenching bath, provided that the temperature was such as to give a completely martensitic initial structure. The results obtained by oil quenching or air cooling, which were found to be practically identical with those from stepped quenching under the conditions described above, have been incorporated into the same group of curves.

The whole of these results are represented by the curves *M*, which therefore correspond to an initial martensitic state, irrespective of the particular quenching conditions. The two other curves *B* and *P*, on the same graph, refer respectively to stepped quenches in the bainitic and the pearlitic regions. For the first case, the mechanical properties are those corresponding to an intermediate temperature of 300° C., and in the second case to an intermediate temperature of 650° C. These two conditions are the most representative, one because of its completely bainitic structure, the other because of the presence of troostite.

The different values of tensile strength, plotted as abscissæ, are those obtained, all other conditions being the same, by tempering at different temperatures. The values of the mechanical properties obtained by furnace cooling after reheating at 850° C. are also shown on these graphs.

The conclusions to be drawn from the graphs are as follows.

The highest mechanical properties are realized, whatever tensile strength is required, by stepped or direct quenching, so as to form an initial structure which is entirely martensitic. For tensile strengths greater than 140 kg./sq. mm. a stepped quench at 300° C. gives, in the bainitic region, a better tensile strength but an appreciably lower elastic limit. The same property was found in the case of the chromium–molybdenum steel studied in our previous paper.

Stepped quenches carried out in the bainitic region lead to mechanical properties equal in all respects to those obtained by martensitic quenching in the range of tensile strengths from 90 to 110 kg./sq. mm. It is to be noted that this range of tensile strength is obtained by tempering at high temperatures, about 600–700°. For tensiles greater than 140 kg./sq. mm., the impact strength, as already shown, is higher, but the elastic limit is lower than the values obtained by martensitic quenching. Finally, in the range between 110 and 140 kg./sq. mm., all the mechanical properties resulting from a bainitic initial state are appreciably inferior to those resulting from a martensitic initial state.

A stepped quench at 650° with a period of more than about 2 hr. isothermal heat-treatment gives, in every case, mechanical properties definitely inferior to those obtained by other heat-treatments.

During furnace cooling, a bainitic structure is formed, similar to that obtained by isothermal heat-treatment at 350° C. The furnace-cooled and untempered metal is softer than that quenched with isothermal treatment at an intermediate temperature of 300° or 350° C. and tempered at

TABLE II—*Heat-Treatments of a Nickel–Chromium–Molybdenum Steel Leading to the Highest Mechanical Properties*

Tensile Strength, kg./sq. mm.	Heat-Treatment	Remarks
90–110	1. Direct quench in air or oil, followed by tempering at 600–700°. 2. Stepped quench with isothermal holding between 200° and 700°, followed by tempering at 600–700°.	Equivalent mechanical properties
110–140	1. Direct quench in oil or air, followed by tempering at 550–600°. 2. Stepped quench with isothermal holding at 200° or between 400° and 700°, followed by tempering at 550–600°.	Equivalent mechanical properties
Greater than 140	1. Direct quench in oil or air, followed by tempering at 200–550°. 2. Stepped quench with isothermal holding at 200° or between 400° and 700°, followed by tempering at 200–550°. 3. Stepped quench with isothermal holding between 250° and 350°, followed by tempering at 200° and 450°.	Equivalent mechanical properties Higher impact strength, lower elastic limit

N.B.—For stepped quenches between 600° and 700° C., the period of isothermal treatment should in no circumstances exceed 2 hr.

200° C. ; its elongation at fracture is greater, but its impact strength is slightly less. The two heat-treatments lead to similar values of elastic limit and of reduction of area.

These observations, which are on the whole very similar to those made on the chromium–molybdenum steels, may be assembled in the form of a Table showing the heat-treatments which lead to the highest mechanical properties (*see* Table II).

IV—Summary and Conclusions

The essential aim of this investigation was the study of the mechanical properties of a nickel–chromium–molybdenum structural steel containing about 0·3% of C, 2·5% of Ni, 1·5% of Cr, and 0·5% of Mo, after different heat-treatments and, in particular, after stepped quenching with isothermal heat-treatment in salt or lead baths for different times and temperatures, followed by tempering at different temperatures.

Firstly, the conditions of isothermal decomposition of austenite were investigated, and compared with those already studied in a previous paper dealing with a chromium–molybdenum steel. The **S** curve shows that different temperature regions may be defined, corresponding respectively to the precipitation of ferrite, the formation of pearlitic constituents, the stability of austenite, the appearance of intermediate structures, and a martensitic transformation. It was found that the presence of the nickel results in a lowering of the transformation points by an amount substantially the same as the lowering of corresponding temperatures on the **S** curve. But the essential action of the nickel is in considerably increasing the stability of the austenite and the critical quenching rate, a phenomenon universally known and confirmed by the trend of the **S** curve.

Examination of the mechanical properties shows their close connection with the conditions of decomposition of the austenite. It is found that so long as the stepped quench gives an initial structure which is entirely martensitic, the mechanical properties are the same, whatever the temperature and duration of the isothermal heat-treatment. At the same time, when samples of small section are used for heat-treatments, direct quenching in oil or air gives results equivalent to those obtained by stepped quenching with isothermal heat-treatment in the region where austenite is stable.

The different mechanical properties have been studied as functions of the temperature of the intermediate bath and the temperature of tempering, and then compared with one another under conditions of equal tensile strength. Three representative states have been considered, *viz.*, martensitic, bainitic, and pearlitic, each one being obtained under particular quenching conditions and corresponding to different mechanical properties. The best results are obtained from samples whose initial structure is entirely martensitic, whether obtained by quenching in oil, in air, or by stepped quenching. Similar results are obtained by stepped quenching with isothermal heat-treatment in the bainitic region, but this holds only for tensile strengths of between 90 and 110 kg./sq. mm. obtained by tempering at a high temperature. Test specimens of bainitic structure tempered at low temperatures and corresponding to a tensile strength above 140 kg./sq. mm., are characterized by a higher impact strength and a lower elastic limit than those whose initial structure is martensitic. Finally, the mechanical properties resulting from a stepped quench in the pearlitic region with isothermal heat-treatment at about 650° for more than 2 hr., are in all cases considerably inferior to those resulting from all other heat-treatments.

In conclusion, the nickel–chromium–molybdenum steel, characterized by a wide region of stable austenite, is particularly suitable for stepped quenching. The range of temperatures at which isothermal heat-treatment produces a martensitic quench is in this case appreciably greater than for the chromium–molybdenum steel without nickel. The choice of temperature for the intermediate bath is therefore less critical.

The mechanical properties obtained by stepped quenching in the region of stable austenite are, as was to be expected, of the same order as those resulting from traditional quenching. On the other hand, the bainitic structures formed by the decomposition of austenite in the intermediate region show distinct characteristics ; the main one being a higher impact strength and a lower elastic limit, except between 110 and 140 kg./sq. mm. This result confirms that already observed in the case of the chromium–molybdenum steel.

It is to be noted that good mechanical properties result from small temperature gradients in the quenching, which appears to constitute the essential advantage of the stepped quench. By this means the consequences of too rapid a cooling rate are avoided ; these are sometimes harmful ; *e.g.*, quenching cracks, deformations, and residual stresses resulting from thermal contractions of varying magnitude during quenching.

It is true that for samples of small section, such as have been investigated in this case, the nickel–chromium–molybdenum steel undergoes a quench to the core simply by exposure to the air. The disadvantages referred to are to be feared less in this case than for samples of larger section which necessitate more forceful means of cooling. It appears that an isothermal heat-treatment in the region of stable austenite is capable of forming

martensitic or bainitic structures even to the centre of specimens of large section, since the temperature must be held isothermally at 650° for at least 2½ hr. in order to start the transformation. It is therefore to be presumed that a stepped quench would lead to mechanical properties at least equivalent to those from traditional methods, for samples of large dimensions. This important point, however, still requires confirmation, and we place it in the front rank of tomorrow's problems.

As is well known, a variety of structures is obtained by isothermal quenching, and the use of these different structures may be questioned. The answer, clearly enough, is that the production of the finest and most homogeneous structure is always desirable. It has long been admitted that martensite quenching is the best procedure for producing these structures, provided it is carried out in such a way as will avoid microfissures and deformation. In this respect isothermal quenching is of unquestioned interest.

When the relations between bainite structures and essential mechanical properties are discussed, it is not always clear which bainite is being considered, for this structure occurs in various forms. Bainite structures generally give results close to those obtained with tempered martensite when the structures are fine, as in the case of the lower bainites. On the other hand, with the coarser medium and higher bainites, the results are not so good.

The results of German researches seem to show that bainite is more heat-resistant (creep-resistant), but here again it would be desirable to define which bainite is referred to, and, so far as possible, to carry out tests with pure and with well-defined structures.

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Note on the Temperature Variation of Young's Modulus of Various Steels*

by F. W. Jones, Ph.D.,† and J. Nortcliffe, B.Sc.†

Synopsis

In a recent paper by Roberts and Nortcliffe it was shown that there is a close similarity in the temperature variation of Young's modulus of different ferritic steels. Additional results have been obtained to broaden the basis of this generalization. Results for austenitic steels, on the other hand, do not show a correspondingly close similarity.

In a recent paper by Roberts and Nortcliffe‡ a description was given of a vibration method for the determination of Young's modulus which is particularly convenient for measurements at high temperatures. Results at temperatures up to 1000° C. were given for several steels. It was noticed subsequently that the variations with temperature (up to 600° C.) of all the ferritic steels were practically identical. Table I gives the experimental values of Young's modulus at 200° C., 400° C., and 600° C. abstracted from the paper by Roberts and Nortcliffe. Figures obtained by multiplying the room-temperature (20° C.) value by a factor f are shown in brackets. The agreement is rather better than might be expected from the estimated experimental error of ± 2%; only in one case (steel F.G. at 600° C.) is the difference between the experimental and calculated value as much as 2%.

Although the steels listed in Table I cover a fair range of chromium contents (F.G. contains 12% of chromium), none of the steels investigated was high in nickel or carbon. It was therefore decided to do measurements on a 4¼% nickel steel and on E.N. 31, which is high in carbon. In addition E.N. 31 may be tempered to a wide range of hardnesses and so is particularly suitable for studying the effect of heat-treatment.

It was considered that if this further work gave the same values of the temperature factor as in Table I, it would be a fairly safe generalization that the factor f could be assumed to be

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‡ M. H. Roberts and J. Nortcliffe : *Journal of the Iron and Steel Institute*, 1947, vol. 157, Nov., pp. 345–348.

TABLE I—*Temperature Variation of the Young's Modulus of Ferritic Steels*

Steel No.	Type	Young's Modulus, tons/sq. in. × 10^3			
		20° C. ($f = 1{\cdot}000$)	200° C. ($f = 0{\cdot}948$)	400° C. ($f = 0{\cdot}875$)	600° C. ($f = 0{\cdot}775$)
1	Mild steel	13·5	12·7 (12·8)	11·8 (11·8)	10·4 (10·5)
2	0·3% Carbon	13·9	13·2 (13·2)	12·2 (12·2)	10·8 (10·8)
3	S.L.V.	13·7	13·1 (13·0)	12·0 (12·0)	10·6 (10·6)
6	2¼% Ni–Cr–Mo	13·2	12·5 (12·5)	11·5 (11·5)	10·3 (10·2)
4	F.G....	14·2	13·4 (13·5)	12·3 (12·4)	10·8 (11·0)

the same for all ferritic steels, independent of composition and heat-treatment.

The analyses of the 4¼% nickel steel and E.N. 31 were as follows :

	C, %	Si, %	Mn, %	Ni, %	Cr, %
4¼% Nickel ...	0·35	0·20	0·41	4·21	1·22
E.N. 31	0·97	0·16	0·38	0·26	1·37

Table II gives the experimental results obtained with these steels. Calculated values are again given in brackets.

The bars of E.N. 31 were oil-quenched after ½ hr. at 825° C. and tempered as follows :

(*a*) At 720° C., to diamond pyramid hardness 260.
(*b*) At 400° C., ,, ,, ,, ,, 600.
(*c*) No tempering, ,, ,, ,, 880.

The agreement between the experimental and calculated values is excellent. Specimen E.N. 31 (*b*) was not heated above 400° C., or further tempering would have taken place. The Young's modulus of E.N. 31 (*c*) at room temperature is merely given to illustrate the effect of hardness on the room-temperature value. Although the hardness of the specimen tempered at 400° C. is much nearer to that of the as-quenched specimen, its Young's modulus at room temperature has increased almost to the value of the specimen tempered at 700° C.

Although the temperature variation of Young's modulus is the same for all the ferritic steels investigated, considerable variations occur in the room-temperature value, which do not appear to be explicable in terms of chemical composition. To obtain some idea of what variations might be expected from cast to cast of steel made to the same specification, measurements were made at room temperature on six steels from different casts, the analyses of which were within the following ranges :

C, %	Si, %	Mn, %	Ni, %	Cr, %
0·28–0·36	0·27–0·34	0·63–0·78	3·19–3·40	0·11–0·29

The values obtained were from 13·2 to 13·3 thousand tons/sq. in., so that the variations in the room-temperature values given by Roberts and Nortcliffe cannot be explained by a cast-to-cast variation. Unfortunately no record was available of the method of preparation of the sample ; it is possible that in some cases there may have been some preferential orientation in the bars.

TABLE II—*Temperature Variation of the Young's Modulus for Ferritic Steels : Further Experimental Values*

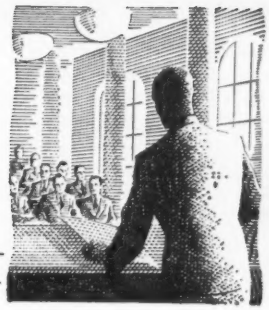
Type	Young's Modulus, tons/sq. in. × 10^3			
	20° C.	200° C.	400° C.	600° C.
4½% Nickel	13·2	12·5 (12·5)	11·5 (11·5)	10·2 (10·2)
E.N. 31 (*a*)	13·8	13·1 (13·1)	12·2 (12·1)	10·8 (10·7)
E.N. 31 (*b*)	13·6	12·8 (12·9)	11·8 (11·9)	...
E.N. 31 (*c*)	13·1	...	...	...

The results for the austenitic steels reported by Roberts and Nortcliffe are less susceptible to generalization. Steels I 6 and Vickro have high nickel contents and the Young's modulus shows a rather smaller variation with temperature than the remaining steels investigated. If we take the following values of the temperature factor f :

Temperature, ° C.	20	200	400	600	800
Factor, f ...	1·000	0·930	0·845	0·762	0·680

The differences between calculated and experimental values for the steels F.D.P., R. 326F, F.V.S., and R. 78 are found to be up to ± 5% at 800° C., although correspondingly less at lower temperatures.

DISCUSSION ON PAPERS



DISCUSSION AT MEETINGS · WRITTEN CONTRIBUTIONS

DISCUSSION ON A
Symposium on Powder Metallurgy

On Wednesday and Thursday, 18th and 19th June, 1947, a meeting organized by the Iron and Steel Institute was held in the Lecture Theatre of the Institution of Civil Engineers, Great George Street, London, S.W.1, to discuss Special Report No. 38—"A Symposium on Powder Metallurgy." A small exhibition of powder-metallurgy products was displayed in an adjacent room during the meeting, and a buffet luncheon was arranged in the Library of the Central Hall, Westminster, on the Thursday.

The Council of the Iron and Steel Institute issued a cordial invitation to all members of the Institute of Metals to attend the meeting, and to the Institution of Mechanical Engineers, the Institution of Electrical Engineers, the Institution of Production Engineers, the Institution of Automobile Engineers, and the Society of Chemical Industry, asking them to inform their members of the meeting and extending to them a cordial welcome to be present.

At the AFTERNOON SESSION on Wednesday, 18th June, from 2.15 P.M. to 5.0 P.M., Mr. D. A. Oliver, in the unavoidable absence of Dr. C. H. Desch, F.R.S., President of the Iron and Steel Institute, occupied the Chair and introduced the Symposium. Then followed the presentation of Section A—"Powder Metallurgy in Great Britain," Dr. W. D. Jones, acting as rapporteur; discussion on Section B—"Preparation, Properties, and Testing of Metal Powders," Dr. L. Northcott acting as rapporteur; and discussion on Section C—"Magnetic Powders and Products," Dr. L. B. Pfeil acting as rapporteur.

At the MORNING SESSION on Thursday, 19th June, from 10 A.M. to 12.45 P.M., Colonel P. G. J. Gueterbock, C.B., D.S.O., M.C., President of the Institute of Metals, was in the Chair. Discussions were based on Section D—"Hard-Metal Carbides," with Dr. C. Sykes, F.R.S., as rapporteur, and on Section E—"Porous Metal Components," with Dr. W. D. Jones as rapporteur.

At the AFTERNOON SESSION on Thursday, 19th June, from 2.15 P.M. to 5.0 P.M., Lt.-Col. Lord Dudley Gordon, D.S.O., President of the Institution of Mechanical Engineers, occupied the Chair. Section F was discussed from two aspects: "The Manufacture and Properties of Sintered Ferrous Components," based on Papers 18 to 22, 27, and 28, Dr. L. Northcott acting as rapporteur; and "The Manufacture and Properties of Sintered Non-Ferrous Components," based on Papers 23 to 27, Dr. Maurice Cook acting as rapporteur.

INTRODUCTION

The **Chairman** (Mr. D. A. Oliver), introducing the Symposium, said: My first duty is to apologize for the absence of our President, Dr. C. H. Desch, F.R.S., who has been unfortunately prevented from attending this afternoon. I have been asked to take the Chair, on behalf of the Institute.

I should like to point out that this Symposium is by no means a parochial affair. In planning it, we have sought the help of the Institute of Metals and other institutions. Besides co-operating with the Institute of Metals, we have invited the help of the Institution of Mechanical Engineers, and invitations have also been extended to the Institution of Electrical Engineers, the Institution of Production Engineers, the Institution of Automobile Engineers, and the Society of Chemical Industry.

A very good history of what has led up to this Symposium is given at the beginning of the Report, so I do not propose to occupy your time in dealing with that subject.

I will now call upon Dr. W. D. Jones to present Section A, entitled "Powder Metallurgy in Great

Britain." As you know, Dr. Jones is the author of the most famous British textbook on the subject. He combines academic thoroughness with practical skill and judgement. For some years now he has been a consulting metallurgist, and from his lectures up and down the country, you all know that he is an authority on the subject. We are therefore very glad that he is to give us this initial historical survey.

Discussion on

Section A—POWDER METALLURGY IN GREAT BRITAIN

and

Section B—PREPARATION, PROPERTIES, AND TESTING OF METAL POWDERS

Dr. **W. D. Jones** (London) (rapporteur), in presenting Section A of the Symposium, said : It is both a pleasure and a privilege to me to present this paper introducing the first symposium on powder metallurgy in this country. I think it is only right and proper that we should have a symposium on powder metallurgy in Great Britain and in fact in London, because it was in London that the birth of the modern industrial use of the technique took place.

Dr. **L. Northcott** (Armament Research Department, Ministry of Supply) (rapporteur), in the presentation of Section B of the Symposium, said : I should perhaps apologize to the authors of the four papers on which I have to report for the omissions which I am now about to make in their papers. I will deal with them in the order in which they appear in the Symposium.

The first paper is by Dr. G. E. Gardam, on the " Production of Iron Powder by Electrodeposition." The major part of the paper is taken up with a determination of the experimental conditions to provide a loose powdery type of deposit. The author starts with a consideration of the conditions at the cathode and then goes on to consider the particular difficulties which occur in the case of iron, such as its reactivity in the presence of air and water, the rapid hydrolysis of ferrous and ferric salts, and the rapid solution of the anodes relative to the cathode current efficiency, leading to the need for including insoluble anodes in the circuit. The author then describes briefly a small experimental pilot plant, capable of making about ¼ lb. of iron powder per hour. This plant consists of a rubber-lined tank, fitted with anodes, half of which are made of mild steel and half of lead. The cathodes are nickel-plated brass tubes, and the powder is scraped from them every 2 hr. The author follows that with a fairly detailed description of the properties of the powder. The powder as produced is not very suitable for compacting, owing to its oxygen content, but Dr. Gardam includes a table showing the improvement on annealing in hydrogen at a temperature of about 700° C. for 1 hr. That is followed by a brief comment on the preparation of hard, brittle electrolytic powder and by a further brief comment on the electrolytic reduction of ferrous hydroxide.

The second paper, by Dr. Miller, on the " Production of Ferrous and Non-Ferrous Metal Powders," summarizes the different methods of preparation which are available. The author classes them into eleven groups, the eleventh being a variety of miscellaneous methods. The first one is milling, and the author refers to the various types of mills, such as ball mills, disc crushers, and so on. The milling process for making powders is of importance, not only because it is an adjunct to other methods of breaking up metals, but because it was used in Germany during the war to a very considerable extent for making large quantities of iron powders. The second group is machining, which is used to a comparatively small extent in this country, and the next is atomization, which is in effect applicable to all metals or alloys which can be melted by running the molten metal through a nozzle or jet. This method was developed considerably in Germany during the war for the manufacture of iron powders. The next group is granulation, and the next is graining, and the author then goes on to discuss the reduction of metallic oxides. This process has the advantage that the metallic oxides can be readily crushed to a small size, and the process results in a very fine metallic powder—1μ or even smaller in size. The next process is the reduction of metallic salts. There are certain metals whose oxides do not readily reduce, and the author gives the instance of the reduction of potassium tantalum fluoride by heating it with metallic sodium to give metallic tantalum. The next process is the electrolytic process, which has been discussed already to some extent in Dr. Gardam's paper, making either a soft powder type of deposit or a brittle one which is subsequently powdered. The next two processes are the carbonyl process and the hydride process, which is particularly used for metals that form hydrides. In the eleventh group the author includes sundry methods which presumably he considers of insufficient importance to warrant a separate classification, such as chemical precipitation.

The third paper is on " Some Properties of Engineering Iron Powders," by investigators from the Armament Research Department. These investigators found in the early stages of their work that the information available on the processes and apparatus for determining the properties of powders was relatively scanty, and they found it necessary to undertake a fairly detailed investigation of the methods of testing. Part I of the paper deals with the methods of testing the various properties. In Part II of the

paper the authors use these methods for examining 28 different iron powders, half of them from American sources and half from British. Most of the powders were either electrolytic or oxide-reduced, but three other types of powders—carbonyl, abrasion, and chloride-reduced—were also included. By a study of those powders under the standardized conditions a large range of properties of the different powders was found. Amongst the properties were those of sintered compacts prepared by similar pressing and sintering techniques. The test compacts were prepared from each powder, after compacting at 30 tons/sq. in. and sintering for 1 hr. at 1050° C. in a hydrogen atmosphere, and were examined, and again a large range of properties was found. Part IV of the paper contains a discussion of all the results, and it includes a correlation of the different properties, with a view to determining which properties are of particular significance. The paper has a useful Appendix by Dr. F. M. Lea, of the Building Research Station, on a modification of the air-permeability method of determining the specific surface of iron powders.

The fourth and last paper is by Dr. H. Meyersberg, on flake metal powders. He starts with a comparison of flake powders with other powders used in powder metallurgy, the essential property of flake powders being that the individual particles are in the form of flakes having a ratio of length to thickness of between 60 and 100 to 1. The author then discusses the application and the use of flake powders. They are of importance in the paint, plastic, printing, and paper-processing industries. (The principal metals used for flake powders are copper, bronze, and aluminium—copper and bronze principally for decorative effect and aluminium for its resistance to corrosion and for certain other properties.) Dr. Meyersberg then deals with the methods of producing flake metal powders, the first being the stamping process. In order to obtain a suitable flatness of the powder, *i.e.*, a flake, it is necessary to subject the raw material, in the form of splashings, to pressure or impact, and stamp mills are used for that purpose. The author briefly traces the development of stamp mills from the early six-rammer open type to the modern variety. In order to obtain suitable flattening, in the first instance it is necessary to use fairly heavy stamps, but, in order to avoid excessive breaking up of the powder, later stamping has to be done with stamps of lighter weight. Normally three types are used, heavy, medium, and light. The author illustrates examples of these, an 18-rammer stamp in the heavy type and a 28-rammer stamp in the light variety. After stamping it is necessary to polish the powders, and that is done in a polishing drum, which consists of a rotating drum with brushes inside, the powder with a small quantity of lubricant being fed into the drum and either the brushes or the drum or both being rotated so that the individual flakes are suitably polished. One disadvantage of the stamp-mill process is the lack of control of the oxidation, which in the case of aluminium powders in particular is important, as it gives rise to severe explosion risk. In order to avoid this explosion risk, the ball-mill process is used. The machine is a modified ball mill, which is described by Dr. Meyersberg as a flattening drum. The drum is fitted with horizontal slats which carry the balls and the raw material about three-quarters of the way up the drum, then they fall and the powder is flattened. By the use of a reducing controlled atmosphere not free from oxygen, since there is a certain degree of oxidation in the case of aluminium (oxygen contents of the order of 3–5% are required), and by there being a sufficiently high velocity in the atmosphere, the fines can be blown through the flattening drum into a cyclone and from there into the polishing drum.

In this section of the Symposium, two papers give experimental and investigational results, one is a review of the methods of production, and the fourth is a description of the methods of making flake powders.

Dr. **J. C. Chaston** (Messrs. Johnson, Matthey and Co., Ltd., London) : I feel honoured by being asked to open this discussion, but I must confess to a certain dismay at having to comment on so distinguished a set of papers.

Dr. Jones has asked for additions to his review, and there is one minior point which I might mention. He says that the manufacture of platinum by powder metallurgy went out of use in 1859 or 1860. I do not think that is true ; and although not all platinum has been made in that way, a certain proportion of platinum has been made in this country by the Wollaston process continuously from its inception until the present day. As far as I know, the equipment and the methods used have changed very little. I do not know whether that is a confession of lack of initiative or whether it is a tribute to the original process.

With regard to Dr. Gardam's paper on the preparation, properties, and testing of metal powders, it should be pointed out that the manufacture of iron powder by milling, that is, by first depositing a brittle sheet and then grinding it to a powder in a ball mill, has been practised since about 1921. I think some reference ought to be made to a paper by B. Speed and G. W. Elmen,* published in 1921, in which are described fully the methods used by the Western Electric Company of those days in making iron powder. Incidentally, that method of making iron powder was the first method used in the manufacture of loading coils. I think the original paper gave all the details of the electrolyte used, and more details were given in a paper published in the *Journal of the Electrodepositors' Technical Society* in this country. An interesting point is that the shape of the particles made by that process is very similar to that of the long needle-shaped particles shown in Fig. 2 of Dr. Gardam's paper. If brittle sheet is deposited, long columnar crystals grow out from the surface of the cathode. Dr. Gardam sweeps them off almost as fast as they are formed, and he does the electrodeposition more slowly, but otherwise his process does not differ in essentials from the

* *Transactions of the American Institute of Electrical Engineers*, 1921, vol. 40, pp. 1321–1359.

older process which was operated successfully for many years.

I do not think that Dr. Miller, in reviewing processes of powder manufacture, has omitted any major method, but personally I should like him to have given more detailed information of the methods used. In the next paper, by members of the Armament Research Department, details of the properties of an enormous number of powders are given, and it would be useful to be able to correlate those with the methods by which the powders are made. It is a pity that we could not have in this paper details of the Hametag iron powders made in Germany, as these might behave rather differently from the powders mentioned in the paper.

If, as a result of this Symposium, those engaged in manufacturing and those engaged in testing could get together and give more information about how the powders are made, as well as the properties of the powders, it would do much to help towards a better understanding of powder metallurgy.

Dr. **W. D. Jones** (London) : There is a good deal that I should like to say about the paper by Dr. Northcott, Mr. Leadbeater, and Mr. Hargreaves, which I think is one of the most important papers that have been published on metal powders. It may be regarded, I think, as a classical study on the subject.

The work is very largely a survey which leads to generalizations about metal powders. The authors say a good deal about the influence of testing methods upon the product, and I think their work shows the value of standardizing testing procedure for metal powders—density tests and flow tests—and the need for a much closer examination of the various methods than there has been in the past. As a matter of fact, it will be seen from this Symposium that various authors use various methods for testing their powders, with different results, and, in fact, they go further than that for they even use different words for the same thing. The American Society for Testing Materials has formed a Sub-Committee which is attempting to produce an agreed glossary of terms used in powder metallurgy, I think with some success, and is going on to standardize powder-metallurgy testing methods. I think that one of the matters to which the Iron and Steel Institute might give attention in the future is the question whether we in this country should adopt the American definitions and testing methods.

As I have said, the paper consists very largely of statistical results and generalizations. I happen to be more interested in the particular than in the general, and, in reading through the paper, I have noticed iron powder No. 25. I know it is no use asking Dr. Northcott where it came from ; in fact, I did ask him and he would not tell me. Powder No. 25 produced compacts with the highest tensile strength, elongation, and hardness, it was one of the highest in impurity content generally, and it had the highest oxygen content, 1·18%, which is higher than the oxygen content of Dr. Gardam's electrolytic-iron powder, which he said was practically useless because the oxygen content was so high. That is one of the contradictions in this Symposium. Moreover, the authors say that any treatment they gave to this powder, including the reduction of that oxygen content, failed to improve it, and I should be glad if the authors would tell us why that was so.

Dr. **Maurice Cook** (Imperial Chemical Industries, Ltd., Metals Division, Birmingham) : In the important and interesting paper by Dr. Northcott, Mr. Leadbeater, and Mr. Hargreaves, considerable stress is very rightly laid on the flow test, and whilst I have no particular comments to make on the work which the authors have done in determining flow values, I would emphasize the very great need for and importance of establishing a standard test of this kind. In powder metallurgy, particularly in those applications where components are made in large quantities at high speed, the question of flow value is obviously a very important one, and the work which Dr. Northcott and his collaborators have done so far will be useful and helpful in formulating and establishing a standard flow test. For applications of this kind it may not be necessary to make the test as elaborate as that suggested in the paper, for since the powder must have very good flow properties there would be no point in having a wide range of flow values.

On the matter of terminology there are two points which might be mentioned. Dr. Northcott and his colleagues, in describing the orifice at the base of the cup, refer to its radius, and in the tests which they carried out on the effect of the length of the stem, which is a tube at the base of the cup, they also refer to the radius ; but the diameter would perhaps be more usual and appropriate in both instances, for the word " radius " is not usually applied to tubes.

Dr. Pfeil, in his very informative contribution on carbonyl-iron powder, in Section C of the Symposium, says that carbonyl-iron powder has a finely crystalline " onion " structure. Whilst this clearly indicates the appearance of the structure, it would again in the interest of suitable terminology be preferable to avoid—when there are adequate alternatives—the use of words of this kind in referring to metallurgical products, and the word " concentric " would seem to be quite appropriate for this particular structure.

Dr. **J. C. Chaston** (Messrs. Johnson, Matthey and Co., Ltd., London) : I should like to refer to the question of taking photographs of metal powders. I am sure that we get an enormous amount of information from the appearance of metal powders under the microscope, and I should like to know how other people photograph them. Dr. Northcott and his colleagues suggest supporting the powders on a stainless-steel plate. I have had the best results by distributing the dry powder on a glass slide by gentle tapping, and illuminating it partly by oblique illumination from a ring illuminator and partly by transmitted illumination from a substage mirror. A combination of obliquely vertical and transmitted illumination can thus be obtained which gives a very fine appearance of depth, and I believe that if that method could be developed it would probably provide us with the best way of seeing the particles.

The photographs which are reproduced in Dr. Northcott's paper are very fine, but I think they would have been very much better if an objective with a greater depth of focus had been used. There is a real need in powder metallurgy for such objectives so that one can see all round the particles.

Dr. **W. F. Chubb** (London): I am particularly interested in the paper presented by Dr. Miller, because he and I collaborated together at Murex, Ltd., in producing tungsten powders, and also because there is a very notable omission from his paper, an omission to which I would like to refer. In 1930, at the Battersea Polytechnic, I experimented with a purely chemical method for producing iron powder. The procedure is fundamentally very simple and is one which in my view may have an enormous practical application in the future. If the iron is dissolved to produce an acetate solution, one can, by neutralizing very carefully and adding certain organic reducing agents, precipitate the iron in a very finely divided form. The method depends upon maintaining the appropriate conditions for the reduction, and provided that these conditions are attained it is possible to secure almost theoretical yields of a very finely divided metal in a state of extremely high purity. My experimental observations also show that other metals may be similarly precipitated from their solutions, or by the decomposition of certain of their derivatives, by using either an essentially similar type of reaction or suitable modifications thereof. While Dr. Miller's omission to mention these possible methods is excusable on the grounds that no details have as yet been published, it is my considered view that these basically simple reactions are capable of development on a production scale. They would appear, indeed, to provide the only means by which metals may be obtained on a large scale in a condition of very high purity and of uniform and controllable particle size.

Dr. **L. Northcott** (Armament Research Department, Woolwich): There is one comment that I should like to make on Dr. Miller's paper. With regard to his fifth process, graining, he says: "The particles are irregular in shape and contain appreciable amounts of oxide, and are not suitable for powder-metallurgy purposes"; I should therefore prefer that method to be included in the miscellaneous section at the end of the paper. In that section Dr. Miller says: "Zinc powder is sometimes produced by atomization but large quantities of very fine powder are produced by the controlled condensation of zinc vapours"; hence I should prefer the condensation method of making powders to be given a somewhat higher status. I do not think it should be classified under the miscellaneous section with a number of other processes which are not, strictly speaking, thoroughbred metallurgical processes.

Under the heading "Reduction of Metallic Salts" Dr. Miller says: "The final product is a spongy material which can be crumbled to produce any size of powder required, down to the low-μ range; this is a particularly useful feature when free-flowing powders are required." That appears to indicate that by getting down to a low-μ range one increases the flowability. That may be so with this particular method, but I should like to ask Dr. Miller whether it is so. My own experience is that free-flowing powders are not generally of the low-μ range.

Dr. Chaston wants to have details of the Hametag powders. As the work described in our paper was carried out during the war, we were in some difficulty in obtaining such powders, and for that reason we did not include in this classification what we know about them now.

I was very pleased to hear Dr. Jones' remarks on the need for standardization of testing methods, and I would refer to the section headed "Conclusions" at the end of our paper, where we make a plea for standardization and say: "It has been shown that suitable methods and apparatus for the examination of all varieties of iron powders require to be standardized and it is recommended that this aspect of the testing of metal powders be considered and undertaken by interested bodies."

With regard to iron powder No. 25, Dr. Jones was in error when he said that the reduction of the oxide content made this powder worse. The main features about iron powder No. 25 are the small median size, 6 μ, and the fact that the surface must obviously be pure and clean, and free from contamination. Those two points linked together will, I think, probably give the answer. Reference should be made to Table XV of the paper, which gives details of a normal powder with an ultimate tensile strength, as received, of 10·31 tons/sq. in. That powder was split up into seven fractions, ranging from a median size of 65 μ to a median size of 5 μ, and it will be seen that the tensile strength increases from 7·9 tons/sq. in. to 17·7 tons/sq. in. There is obviously a very considerable jump in the mechanical properties when the median size is about 12 μ, provided always that the surface of the powder is clean, so that satisfactory cohesion between the particles is obtained. The peculiarity of iron powder No. 25, to which I might draw attention, is that the oxide content of something over 1% was the same after an attempt had been made to reduce it in hydrogen, and the point there is that this fine 6-μ powder consists of small particles, clean on the outside, with a core of oxide. It was an oxide-reduced powder, and in simple hydrogen reduction the internal oxide was not readily reduced.

Mr. **R. Titterington** (Murex, Ltd., Rainham, Essex): Dr. Miller regrets that it is impossible for him to be present today as he is on business in the United States. At his request I have undertaken to try and answer any points raised in connection with his paper.

There has been a certain amount of criticism of Dr. Miller's paper on the ground that the methods of producing metal powders are referred to without any great detail. As you are aware, when papers were first called for they were limited in size to about 3000 words, and to refer to all these methods in detail was obviously impossible in the limited space available. There is a considerable amount of additional information which could have been published if the circumstances had been different.

The question of microphotography in connection with metal powders was raised by Dr. Chaston, and

in Dr. Miller's paper there are several microphotographs of metal powders. His Fig. 1 shows a milled chromium powder viewed by transmitted illumination, and Fig. 2 shows a sample of atomized aluminium powder viewed by oblique illumination. This method is suitable for relatively coarse powders and gives the indications of particle shape to which Dr. Chaston referred. As will be seen, there is a certain amount of flare in this photograph, *i.e.*, scattered light from the larger particles. In Fig. 3 an aluminium powder of the same type is shown, viewed by transmitted illumination. I think the difference brought out by these two photographs bears out the remarks made by Dr. Chaston in regard to the necessity for further investigation into this problem.

Dr. Northcott raised a point about iron powder having good flow properties where the particle size was in the low-μ range. The heading under which this appears is "Reduction of Metallic Salts." A considerable amount of work was done on the production of very pure iron powders by the reduction of ferrous chloride by hydrogen. The ferrous chloride was obtained by the evaporation of hydrochloric acid pickle liquors and was reduced in hydrogen at low temperatures. The material so produced is a very light sponge, and the μ particle size is very fine. The milling of this sponge does not break down the powder to its ultimate grain-size, but leaves small aggregates the size of which can be varied by the choice of suitable milling conditions. The result is a powder having a very fine true particle size, but in the same form as if it has been subjected to granulation. In this way very good flow properties can be achieved while still retaining the extremely fine true grain-size.

In the paper by Dr. Gardam there is a reference to the direct estimation of iron in iron powders, and he states that his routine determination of the total iron content gives him an accuracy of $\pm 0{\cdot}05\%$. That seems a very high standard of accuracy for routine determination in the case of a metal like iron, and I should like to know how it is obtained.

Mr. **A. G. Metcalfe** (Queens' College, Cambridge) : In connection with Mr Titterington's last remark, it may be of interest to mention that X-ray photographs have shown that the -200-mesh and the $+30$-mesh fractions of hydrogen-reduced copper powder have about the same grain-size. This is around 5 μ, so that these powders must consist of sintered aggregates. The aggregates may be broken down in many of the processes used in powder metallurgy so that the grain-size revealed by X-rays would appear to be more fundamental.

Mr. **V. T. Morgan** (Messrs. Bound Brook Bearings (G.B.), Ltd., Birch Road, Witton, Birmingham) : I would like to refer to the paper by Dr. Northcott, Mr. Leadbeater, and Mr. Hargreaves.

The density of the powders which the authors examined was determined by adding the powder a little at a time into the flow apparatus so that the cup was filled in a series of small quantities. Experiments which we have done have shown that better reproducibility of the density figure can be obtained if the cup is filled by a continuously flowing stream of powder. In the tapped-density measurements, the authors worked with $1\frac{1}{2}$ in. fall. Experiments carried out to determine the number of taps and the interval and severity of tapping, showed that if a fall of 1 in. was exceeded a considerable recoil occurred within the column of powder. We found it better to give a large number of taps with a short fall, and finally we standardized a fall of $\frac{1}{4}$ in. as this seemed to give optimum reproducibility. There is always a density gradient throughout the column, and with large distances of fall the top layers never appeared to compact at all. With small falls the thickness of this top layer is reduced to a minimum and does not appreciably affect the result.

In Fig. 5 of the paper, in which the period of flow is plotted against the length of stem, am I right in concluding that the authors found that the longer the stem the better the apparent flow ?

Dr. **Northcott** : The longer the stem, the shorter the period of flow.

Mr. **Morgan** : That has not been our experience in the case of copper powders. We find that with long stems, flow is more difficult. In Table I the authors draw a comparison between the apparent densities produced by different radii of flow-meters, the figures given being 2·766 g./c.c. and 2·745 g./c.c. A difference of 0·02 g./c.c. is under a 1% difference. Operators who are measuring density and flow regularly day after day would consider that these two results alone are a very good check, and that anyone quoting density figures nearer than 0·02 is splitting hairs. Are the apparent densities shown in Table I single results or the mean of a number of results ? Can they be reproduced ? If so, there is something wrong with the way in which we are doing density determinations. The apparent density is given to four figures. Is that reproducible time after time ?

Dr. **Northcott** : Yes, it is reproducible ; 2·766 is given as the mean of three results.

Mr. **Morgan** : What would be the likely variation for one powder ; is it in the limits of four-figure accuracy ?

Dr. **Northcott** : In the works a powder which gave an apparent density of 2·382 would not be considered vastly different from one which gave 2·766, but in a scientific investigation, I feel that unless we have standards at least reasonably high there is not much point in doing the work at all. I feel that the results which we have shown in Table I indicate that we have a simple test for the determination of powder density which one can carry out on the powder. The results show that even in determining apparent density, a property one would imagine to be simplest of all to determine accurately on a powder, one may get measurable variations unless the conditions are standardized. The figures required will be included in our written reply.

Mr. **Morgan** : It has always been my idea that, in quoting scientific results, one should not quote them to any more accuracy than that to which one can reproduce them. The point I wanted to raise was that if you have a method and a technique of reproducing density figures to four figures, I for one should

very much like to know the details, because 1% accuracy is as much as I have been able to obtain.

The **Chairman** (Mr. D. A. Oliver) : There is an opportunity now for those who have had experience in measuring powders to comment on this very important and by no means sealed and settled question of standardization.

Dr. **E. C. Rollason** (Murex Welding, Ltd., Waltham Cross) : I should like to make one point arising out of the methods of measuring particle size and surface area. Under the heading of " Specific Surface " Dr. Northcott and his colleagues review various methods, but the one selected is that developed by Lea and Nurse, which depends upon air permeability. For another purpose the company with which I am associated developed a very quick method of measuring surface, using the Spekker absorptiometer. The powder is suspended in a suitable liquid, such as ethylene glycol, with the addition of a deflocculating agent. The surface area is estimated from the observation of the optical density of the dilute solution. A description of the method, which is rapid and uses equipment available in most analytical laboratories, has been published.* We ourselves have found it extremely useful for controlling the particle size of various minerals and metal powders; the metal powders, incidentally, have not been used for powder metallurgy. I should like to ask whether this method has been considered by Dr. Northcott and his colleagues for controlling iron powder and other powders. It is very simple and is suitable down to 2 μ, at which fineness the air permeability begins to suffer also from certain defects. I should be glad if the authors could comment on the use of this method.

The **Chairman** (Mr. D. A. Oliver) : I was very interested in Dr. Chubb's contribution with regard to chemical methods of making powders. I was not quite clear whether he said he had published his results or not. I believe that work on very similar lines is being done in different parts of Europe. If Dr. Chubb would give us a historical note of his earlier work in writing later on, I am sure it would be most interesting.

I think it is fair to mention that many of these papers were produced at rather short notice and under great pressure, by a large number of experts in the field, with the severe limitation on length which has been mentioned in the discussion. The Institute has done a magnificent work in getting them through the press in record time, and I think the fact that the Symposium is as balanced as it is, is remarkable.

AUTHORS' REPLIES

Dr. **G. E. Gardam** wrote in reply : Whilst we were aware of the two previous papers on the production of iron powders by grinding of brittle electrodeposited iron, the original object of the present work was rather to obtain iron powder direct, and if possible in a soft and compactable condition. Mr. Chubb's method of chemical reduction of iron salts is of considerable interest, but the problem of efficient washing and drying of the powder without undue oxidation still remains.

The estimation of the total iron content was done by normal volumetric methods by skilled personnel using every care, and in this sense was not a routine determination, but it was considerably more simple than an oxygen determination.

Dr. **L. Northcott,** Mr. **C. J. Leadbeater,** and Mr. **F. Hargreaves** wrote : In reply to Mr. Morgan's remarks, an exchange of experiences and views with other investigators prior to our investigation had indicated that widely different methods were in use, so that the best course for our purpose was to examine and assess various methods and present the experimental facts

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TABLE A—*Apparent Densities* (*g./c.c.*)

	Iron				Copper	
	A	*B*	*C*	*D*	*E*	*F*
	1·0218	2·012	2·186	3·135	1·860	2·585
	1·0040	2·030	2·189	3·147	1·866	2·575
	1·0050	1·996	2·195	3·157	1·866	2·579
Max. difference	0·0178	0·034	0·009	0·022	0·006	0·010
Percentage difference	1·7	1·6	0·4	0·7	0·3	0·4

in the hope that one or other of these methods might have been considered by other workers, who would have contributed materially to the discussion.

Our experience with the method criticized had shown that for the majority of powders available for development or production, satisfactory consistency was obtainable for values of apparent density. It would be rash, however, to claim, and for that matter undesirable to be made responsible for claiming, a high degree of consistency in this determination for all powders. Obviously this cannot be the case for metallic powders obtained from such widely different sources and methods and possessing widely differing properties. It is necessary to bear in mind, for instance, that ferrous powders in particular also are liable to become magnetized, a condition militating against achievement of consistency of results in different methods for apparent density. The experimental results of triplicate tests given in Table A indicate the fairly high accuracy obtainable.

In the four values for iron powders the variations are 0·4%, 0·7%, and 1·6%, 1·7%, whilst for the copper powders the variation is 0·4%, 0·3%, which clearly justifies the use of the method. His experience on the determination of tap density, presumably entirely on non-ferrous powders, differed from our own with ferrous powders for which it was established (*see* Table II, page 18) that varying the height of drop did not affect the final result, provided that a sufficient number of drops was allowed in each instance to attain maximum density. It is not clear from Mr. Morgan's remarks whether he has considered this aspect of the problem and it may account for the apparent discrepancies.

Dr. Rollason's work on a photo-electric method for determining size distribution was published after the work described in the paper had been completed. There appears to be great scope for a simple and rapid method for determining size distribution, especially for production control, and photo-sedimentation methods deserve full consideration. For investigation work it is considered that it cannot supply some information which may be required, and therefore requires supplementing by other methods, *e.g.*, in the determination of shape factor within specific size-ranges.

Discussion on
Section C—MAGNETIC POWDERS AND PRODUCTS*

Dr. **L. B. Pfeil** (The Mond Nickel Co., Ltd., Birmingham) (rapporteur), in presenting Section C of the Symposium, said : The papers now to be discussed have been grouped together as Section C, dealing with magnetic powders and products, and it is notable that a variety of problems can be appropriately discussed in this section. We have papers by Mr. Polgreen and Mr. Buckley, concerned with highly specialized problems of magnetic dust cores, which perhaps have little relationship with the usual problems dealt with in powder metallurgy. Then there are papers by Mr. Garvin and Mr. Oliver, which deal respectively with sintered magnets and with pole-pieces, the production of which involves a technique that does not differ fundamentally from that for normal powder metallurgy. The fifth paper in Section C deals with the production and properties of carbonyl-iron powder.

There are a number of common factors in the papers in Section C, which I will summarize. A characteristic of great importance in magnetic products is permeability. The permeabilities of most ferromagnetic materials are several hundred or several thousand times the permeability of air or space, and from that it is clear that the permeability of a compressed powder will be largely influenced by the admixture of even a small proportion of voids or of non-magnetic material.

The relationship between the effective permeability of a compact and the intrinsic permeability of the magnetic material is illustrated by Mr. Buckley in his paper, under the heading " Constancy of Permeability of Dust Cores." His figures indicate that, in the case of such cores, where the effective permeability lies in the range of 10–150, the intrinsic permeability of the alloy has little effect on the permeability of the core, owing to the important part played by the insulation between the particles. It thus appears that particle size and shape, the behaviour of the powder under pressure, and the thickness of the insulating film between the particles, are of more importance than the intrinsic permeability of the alloy.

In the case of high-permeability pole-pieces, produced from pressed and sintered iron powder, Mr. Oliver shows that the permeability of the sintered material, although never quite as high, reaches the same order as that of solid materials of the same general type. The case then is different here, for, although the effect of the particle junctions is still probably playing a considerable part, the inherent properties of the iron are the more important factor. Possibly this is the reason why the densities of the compact do not correlate more exactly with the permeability. It will be of value if more data of this nature are brought to light in the course of the discussion, and it would be of interest to know whether it is possible to separate the two variables, magnetic dilution due to the voids and inclusions, and magnetic-permeability variations due to the chemical impurities, structure, thermal treatment, etc. With further information it might be possible to form a general theory and to link up these cases with an explanation of the permeabilities of dust cores.

Mr. Garvin states in his paper that the remanence

* The sentence beginning in line 30 of page 56 of the first edition of the Symposium should read : " The low-nickel alloys (*LNF*) are characterized by high permeability and an electrical resistivity value of over twice that of the high-nickel alloys and several times the resistivity of iron ; . . . ."

of sintered iron–nickel–aluminium magnets may be 10% lower than that of the regular alloy, and he suggests that this may be due to porosity.

A property of special importance in magnetic applications is high purity ; this refers not only to the absence of small amounts of mechanically admixed impurities, but also to elements likely to affect the magnetic properties adversely, such as sulphur, phosphorus, and oxygen. The first type of impurity affects the magnetic dilution, and the latter, besides affecting the intrinsic permeability, may also increase the hysteresis losses.

I should like now to refer to a few special points in the two papers that deal essentially with dust cores.

Mr. Polgreen provides in Table I of his paper an excellent summary of the applications for which dust cores are required, and both he and Mr. Buckley discuss the significance of the three causes of core losses in terms of the hysteresis-loss coefficient, the eddy-current-loss coefficient, and the viscosity or residual loss. It is pointed out that the relative importance of these losses varies with the frequency and that different types of powder—iron, nickel–iron, nickel–iron–molybdenum, and others—possess varying characteristics. Accordingly, great skill and experience are required to choose the correct product for each type of application.

The progress that has been made in developing improved dust cores of increased efficiency is certainly an outstanding achievement, and I think we should all be proud of what has been done in this country in that connection, but we should remember that there is a fruitful field here for further research. In a number of important applications higher permeability would be of very great value if it could be obtained without increased losses. It is perhaps not altogether clear from these papers whether the search should be directed towards powders having a higher intrinsic permeability or towards a reduction in the amount of magnetic diluent which is used for insulation purposes between the particles. I very much hope that in the course of the discussion this matter will be clarified. Metallurgists who are concerned with the production of magnetic powder need guidance here from physicists, because if a higher intrinsic permeabilty is particularly desired in metal powder for dust cores it may be necessary to accept increased losses. Again, the metallurgist may well be asking himself whether it would be of value if he could produce an iron powder of high purity and good physical form such that it packed well and had a coarsened particle size, or whether he should direct his attention to a nickel–iron alloy powder with good packing characteristics and a finer particle size than has so far been available. In the first case he might argue that higher permeability dust cores could be produced without the complications involved in a final heat-treatment, and in the second case he might argue that such a powder should give cores with reduced eddy-current losses, which would be suitable for very high frequencies, the losses being reduced by virtue of the relatively high electrical resistance of the nickel–iron alloys.

Mr. Oliver outlines the advantages of sintered iron pole-pieces, more particularly in the saving in machining costs, but at the same time he draws attention to the difficulties which are involved in obtaining magnetic properties fully equal to those obtainable in standard materials. The need to eliminate voids and the effects of particle distortion are brought out. Mr. Oliver refers to preliminary experiments which have been carried out on a 4% silicon–iron powder, which alloy should have the advantage of a higher electrical resistivity, thus reducing the eddy-current losses when an alternating magnetic flux is involved. It might be supposed that the use of silicon–iron powder would increase to some degree the sintering difficulties associated with the use of elements which have a high affinity for oxygen, this being a matter to which Mr. Garvin draws attention. The large surface area exposed by powdered metals provides opportunities for the oxidation of highly reactive elements. Mr. Garvin refers to the merits of pre-alloying, emphasizing how this procedure may be used to avoid the formation of a liquid phase during sintering. At the same time, pre-alloying, as in the case of aluminium and cobalt, referred to by Mr. Garvin, may have a useful effect in reducing the tendency for films of alumina to form on the particles. Again, the production of dense green compacts by the appropriate choice of particle size and shape and of moulding conditions should serve to reduce the difficulties inherent in the handling of highly reactive elements in powder metallurgy. In describing the sintering process, Mr. Garvin gives an account of a method with which he has been concerned for preventing the formation of oxide films, a method involving surrounding the articles during sintering by material having a still higher affinity for oxygen than that of the constituents of the compact.

The growing importance of alloys containing substantial amounts of highly reactive elements, such as niobium and titanium, indicates that much further work is desirable on sintering operations under completely non-oxidizing conditions. Mr. Garvin's paper is therefore particularly welcome.

Finally, I have to introduce the paper by myself, which deals with iron powder made by the carbonyl process. Carbonyl-iron powder is used mainly for dust cores by virtue of the unique shape of the particles, the structure of the particles, and the magnetic properties that they possess. The photomicrographs in Figs. 1 to 4 illustrate some of these characteristics. There are two principal varieties of carbonyl-iron powder, the high-carbon hard grade, which is manufactured in this country under the description "*ME* grade" and is used for high-frequency cores, where eddy-current and hysteresis losses must be at a minimum, and the decarburized soft "*MC* grade," which is used where higher permeability is essential.

Certain applications for carbonyl-iron powder other than dust cores are briefly mentioned in the paper. These uses depend on the high purity of the decarburized powder and on the particle size and shape.

I must apologize to the authors of these papers for the very many omissions that I have necessarily made, but I hope that I have drawn attention to a number of points in terms which will help to produce an interesting discussion.

Mr. **W. F. Randall** (Telegraph Construction and Maintenance Co., Ltd.) : In opening the discussion on magnetic powders my comments will be restricted to what are generally known as dust cores used in telecommunications. The technique of manufacture of such cores differs from the general procedure in powder metallurgy in that whereas in the latter the effort is to make from dust a coherent metallic mass, in the case of magnetic dust cores sintering is prevented by insulation of the particles.

Mr. Polgreen's statement that the intrinsic permeability of magnetic oxides is very low, is not borne out by recent research, for certain of the complex ferrites containing zinc, nickel, cadmium, or copper, develop permeabilities as high as 4000. Sendust, the Japanese alloy, is dealt with rather cursorily. Pre-war samples made in Japan gave a permeability of 110, but it is not easy to obtain this figure for these reasons : Firstly it is difficult to make a homogeneous alloy by ordinary melting methods and, secondly, there is a very critical optimum composition. The problem of manufacture might be solved by sintering from powders to get a homogeneous alloy and exact composition.

Dr. Pfeil suggested the discussion might bring out some lines of research. He emphasizes the pre-eminence of carbonyl iron, which is due to the two factors, its particular particle size and particle shape. These are determined by the manner of manufacture, and if it were possible to make a composite nickel iron by the same process we should surely make some great advances. It has been established in Germany that rolled strip made from sintered nickel–iron carbonyl dust gives very much better magnetic properties than melted material.

Mr. Polgreen, in Table II of his paper, does not mention the Hametag mutual-attrition process, as employed in large-scale production. I understand this was widely used in Germany for the production of both iron and nickel–iron powder.

A point which does not appear to be mentioned in the papers on dust cores, but which is referred to in Mr. Garvin's paper, is the possibility of increasing the density of the dust-core compact by grading the powders. The advantage of that process was brought out by Mr. Chesters in a paper read before this Institute. I should like to know whether any further information can be given on that subject.

The method of heat-treatment of these cores should be a profitable line of research. The problem is a difficult one as we have a mass composed of heat-conducting particles each surrounded by a heat-insulating layer, through which heat applied from the outside must be conducted. It would appear that some form of high-frequency heating would be the best solution of the problem.

The photomicrographs shown in Mr. Sinclair's paper, in Section E, were interesting in so far as they show spheroidal particles with almost point contact in grades *A*, *B*, and *C*. If it were possible to maintain some such structure the resistance of the compact should be very high and it might serve as a magnetic dust-core for certain purposes.

Turning to magnetic properties, Mr. Polgreen states that hysteresis loss is dependent upon uniformity of powder. I cannot quite understand to what type of uniformity this refers, as I cannot understand hysteresis being affected by either size or shape.

Magnetic treatment of dust-cores on the same lines as the magnetic treatment of permanent magnets has not been discussed. If it were possible to obtain the magnetic field while the core is being compacted so that all the particles are aligned in the easiest magnetic direction, the hysteresis factor should be considerably reduced.

Figure 1(*a*) of Mr. Polgreen's paper raises an interesting point. The magnetic viscosity of various powders is shown in graphical form, magnetic viscosity being defined by Mr. Polgreen as the hysteresis loss at zero flux density. This viscosity loss is usually at a minimum when the initial permeability is highest, but there are two samples given in Fig. 1(*a*), *CL Fe* and *HNF*, which have initial permeabilities in the ratio of possibly 10 or 20 to 1, yet the *CL Fe* core shows a much lower viscosity.

In Dr. Pfeil's paper there is a discussion on the hysteresis of mechanically hard particles. I wonder whether we are quite right in assuming that low hysteresis is associated with mechanical hardness. When we speak of the hysteresis of these cores we are speaking of the hysteresis factor, which is not quite the same as intrinsic hysteresis. Hysteresis, as we ordinarily understand it, is measured at a given maximum induction while hysteresis in the core is measured at a given magnetizing force. The induction in the hard particles would be very small compared with the induction in the soft particles, so it may well be that the intrinsic hysteresis of the hard particles is actually higher than that of the soft particles.

In the case of these dust cores, the characteristic of the powder which is of the greatest importance after particle size is probably particle shape. The nearer the spheroid the particle approaches, the less the contact between the particles and the lower the eddy-current loss, so that some easy way of determining shape seems to be of great importance. I suggest that if we took size into account and coupled it up with information on air permeability, flow behaviour, and possibly packing density, there should be a mathematical way of determining the shape factor from those data.

Mr. **C. E. Richards** (Post Office Research Station, London) : I should like to join issue with Dr. Cook on the subject of greengrocery. I cannot think of a better way to describe a carbonyl-iron structure than to call it an " onion-skin " structure. That is an expression which has been used for a long time, and it seems to me to describe the structure quite as definitely as does the word " concentric."

There is one slight complaint which I have to make about Dr. Pfeil's paper. In the paragraph

describing the inception of the carbonyl-iron plant, he says that The Mond Nickel Company got down to thinking about it and getting on with it. The Mond Nickel Company did think about it and get on with it, and they made a remarkable job of it. I think the plant must be almost unique ; it was designed by their engineers and it worked without snags from the word " Go." I think it should be recorded that the conception of a carbonyl-iron plant in this country was due largely to the initiative of Mr. Fitzherbert Wright, who was Controller of Iron Powder at the time.

Dr. Pfeil says that the magnetically hard materials are invariably used at low flux densities. Early in the war there was at least one case in which those materials were used at relatively high flux densities. There was one radio set, in which, when an iron powder with a high hysteresis loss was used, the core became red-hot. Carbonyl-iron cores never actually gave trouble on that set.

Mr. Polgreen says in his paper that nickel–iron powder should be better, with regard to eddy-current loss, than carbonyl-iron powder, because it is intrinsically a higher-resistivity material. I thought that was so myself for a long time, but I cannot produce any evidence to show that a 50/50 nickel–iron, which should have a resistivity of approximately 50, is any better with regard to hysteresis loss than carbonyl-iron of the *E* type, whose resistivity I do not think we really know, but we always regard it as 10, which is the normal figure for iron.

On the question of dealing with the losses, Mr. Polgreen uses the Continental system of expressing them, and Mr. Buckley expresses them by the American system. I think it would be useful if we could decide at some time or other, before too much is done, which system we are going to adopt in this country. They are mutually convertible, but it is much easier to have the figures so that one can read them off one against the other. My own preference is for the American system, but that is probably because I started using that system before I came into contact with the Continental one.

In dealing with these figures, I think it might have helped a little if Dr. Pfeil had added some figures for the properties of carbonyl-iron as used in cores. Mr. Buckley quotes in his paper an electrolytic-iron dust with a permeability of 35 and an eddy-current loss of 88. One can press up a carbonyl-iron, grade *C*, to a permeability of 35 quite easily, but, instead of the eddy-current loss being 88, it might perhaps be 2. The hysteresis losses are not very different. I know Mr. Buckley is not suggesting that the iron powder he quotes is equivalent to carbonyl *C*, but the comparison would assist designers.

Mr. Randall mentioned Sendust. It is a chancy kind of material, but the only permeability figures which we have determined ourselves line up far more closely with Mr. Polgreen's figure of 70 than with Mr. Randall's figure of 110. My recollection is that the Japanese Sendust obtained before the war, gave, as a general rule, a permeability of 60–65.

Mr. **S. E. Buckley** (Standard Telephones and Cables, Ltd., London) : I agree with the previous speakers who have said that the dust-core problem has many unique features. Normally, for magnetic nickel–iron alloys in sheet form, it is necessary to have such elements as sulphur kept at a minimum, but for the production of brittle material for dust cores some sulphur must be present, so that there will be a sulphide grain boundary to facilitate the subsequent breaking up of the material. Also it is usually considered that the best magnetic properties are obtained in sheet material when all mechanical strains are eliminated. This is not possible in dust cores, since in order to get high permeability it is necessary to employ forming pressures of 100 tons/sq. in., and thus strain the particles beyond the elastic limit. Hence heat-treatment of the pressed cores is necessary in order to develop the magnetic properties.

As already mentioned by Mr. Randall, the requirements for the insulation of the particles forming the cores are very severe. For permeabilities of 100 and over, the thickness of the layer of insulation has to be about $\frac{1}{2}$ μ. This layer has to stand the pressing operation, and heat-treatment at about 500–600° C. It is well known that the best annealing temperature for strip and sheet is often above 1000° C., but no core-insulating materials which I have met so far will in such small thickness withstand temperatures approaching 1000° C. Again, in other applications of powders the strength of the compact is obtained by sintering, whereas sintering has to be avoided in the heat-treatment of magnetic dust cores. Mr. Randall mentioned the possibility of obtaining an improved heat-treatment of dust cores by the use of high-frequency heating, but it seems that the power required would be very high, because the core has especially been made with the object of minimizing induced eddy-current losses.

Dr. Pfeil raised the point whether the use of high-permeability materials or better methods of insulation are likely to give higher-permeability cores. I think there is scope for both of those methods, but it is possible that there is the greater scope for improved methods of insulation, but the requirements for the insulating layer will be very severe. The possibility of new insulating techniques is important, since at present there is a gap in the range of available magnetic materials for use between 100 and 500 Kc. Thin tape cores of permeability about 1000 yield coils of low *Q*, while the available dust cores for high-*Q* coils are of permeability of only 10–20. New methods of insulation might yield cores of higher permeability and yet of reasonably high *Q* for this frequency range.

Dr. Chaston referred to a pioneer paper by Speed and Elmen on electrolytic-iron dust. This material is referred to in the earlier part of my paper, and was used in cores of permeability 35. Mr. Richards referred to the reduction of losses obtainable by the use of carbonyl-iron as compared with the electrolytic-iron dust of Speed and Elmen, and I quite agree with the values which he quoted.

With regard to Sendust, the value of core permeability which we obtained was about 67, which agrees quite well with Mr. Richards' figures.

As to Mr. Richards' point relating to Mr. Polgreen's

comparison of the eddy-current loss of nickel–iron alloys with that of carbonyl-iron, when we made a comparison between Sendust, which has a resistivity of about 90–100, and some of the nickel-irons, which have a resistivity of about 40–50, and produced cores of the same permeability, made from particles of the same size, we found that the Sendust cores had a lower eddy-current loss.

Another point that has been raised is the question of the calculation of permeability, on which there has been some controversy. Nearly all of the calculations are based on a regular geometrical arrangement of uniformly sized particles, but neither of those conditions obtains in the dust core. In practice not only are there irregular shapes amongst the particles, but the staggering of air-gaps between adjacent layers greatly complicates the computation. A rigorous mathematical treatment involving all of the relevant factors appears to be almost impossible.

With regard to Mr. Randall's comments on Mr. Polgreen's figures relating to variation of viscosity-loss coefficient with permeability, showing that higher values were obtained at higher permeabilities, Mr. Polgreen's measurements are in full agreement with results which we also have obtained independently. I think the explanation is partly as follows. In dust cores of different permeabilities which are pressed at high pressures, the inherent properties of the particles in the different cores are not the same, because in the cores which finally have the higher permeability there is less insulation and less cushioning effect from the insulation, whereas in the low-permeability cores there is considerably more insulation and the strain on the particles during the pressing process is less.

Mr. **A. D. Thomas** (Salford Electrical Instruments, Ltd., Salford, Lancs.) : Mr. Randall referred to Mr. Polgreen's point about magnetic oxides and he mentioned the ferrites. Information on the ferrites was published after this paper was written, and Mr. Polgreen was more concerned with the low-permeability oxides.

The permeability figure of 70, given by Mr. Polgreen for Sendust, was, I think, obtained from German sources, when the Germans were making Sendust to save nickel, and they did not have much success with it.

With regard to the hysteresis loss being dependent upon the uniformity of the powder, it is dependent upon the uniformity of shape as well as on the uniformity of size.

As to the suggestion that the eddy-current ought to be better in the nickel–iron than in the carbonyl-iron, the shape factor comes into the question. A good shape factor also produces a low eddy-current, low loss factor, and better insulation. Some work is being done now on producing a nickel–iron alloy of a similar shape to the carbonyl-iron, and the indications are that the eddy-current loss is lower, but the results are not definitely established yet.

Dr. **J. C. Chaston** (Messrs. Johnson, Matthey and Co., Ltd., London) : I am not at all happy about the suggestion, which has been made several times today, that a low eddy-current loss is necessarily associated with a spherical-shaped particle, on the assumption that such particles touch at only a few points. We have to consider what happens in the pressed core and not in the core before pressing. Adjacent spherical particles which originally touch at only a few points may be very much deformed and conform in their contours over an appreciable area after pressing. I do not think we can look at a particle and say : " This will make a good dust core." The only way to find out whether a powder will make a good dust core is to make a dust core out of it and test it.

I was interested to hear from Mr. Randall that carbonyl-iron–nickel gives such high values of permeability, and I think many people would be glad if Mr. Randall could give a few figures of comparison. It has always been my impression that nickel–iron made by the carbonyl process is in sheet form no better than melted nickel–iron alloys and in powder form is worse.

The **Chairman** (Mr. D. A. Oliver) : I think we are very fortunate in having these two papers by Mr. Polgreen and Mr. Buckley, and we have also had a very good discussion on the insulated particle part of this group.

I should like to comment on Mr. Garvin's paper on permanent magnets, because, from a completely judicial point of view, I believe that the permanent magnets made by powder metallurgy in this country are really better than those made abroad. That underlines the achievements of Mr. Garvin's company and also underlines the value of the newer features of the technique which has been evolved, utilizing not the brittle iron–aluminium but the cobalt–aluminium powders, which, as mentioned in the paper, happen to be more resistant to oxidation.

I should like to ask Mr. Garvin whether he has had any experience of the hydrides, as I understand that the metal hydrides were used to a considerable extent, particularly in America, in the early work on permanent magnets by powder metallurgy.

There is another point on which I think the ordinary powder metallurgist should not be too impatient with the magnetic sub-department, that is, with the advantages of the magnetic method of analysis. We have in it a very wonderful indirect method of measuring the internal stress in the compacted and heat-treated mass, and therefore the effective permeability can be related to the average internal stress, provided that we know what we are dealing with as regards the initial powder.

From a magnetic point of view, iron powders are extremely sensitive, and therefore we have to relate magnetic observations to a tested purity of powder under constant pressing and sintering (or annealing) conditions ; or, conversely, if we have a pure and high-quality powder we can learn much about stress-relief, diffusion, and the mechanism of sintering generally.

With regard to the mysterious magnetic viscosity, I think I am correct in stating that, if we have a perfectly pure iron and it is contaminated with nitrogen atoms, those nitrogen atoms tend to migrate to the grain boundaries and produce a magnetic viscosity which appears as a time lag between the

applied H and the resulting B. It may well be that that factor, which it is so difficult to measure precisely, enters into the questions which have been discussed. It may in fact be nitrogen or minute traces of oxygen, sulphur, or carbon, or any other impurity that is known to segregate at the boundaries or interstitially and so prevent the most perfect annealing from being achieved.

On behalf of the Iron and Steel Institute, I should like to tender our thanks to the Institution of Civil Engineers for the use of this hall, which has added greatly to the comfort of our proceedings.

(The meeting adjourned to the following day, Thursday, 19th June)

AUTHORS' REPLIES

Mr. **G. R. Polgreen** (Salford Electrical Instruments, Ltd.) wrote in reply : Mr. Randall and other speakers referred to my statement that hysteresis loss is dependent upon the uniformity of the powder. This has been proved by experimental work and there would appear to be a straightforward explanation. If a magnetic powder is very irregular in particle shape, when it is insulated or pressed into a core there will be an appreciable local distortion of the magnetic flux due to the large number of high-reluctance points. If, on the other hand, the particles of the powder are made uniform in shape and surface, and have a high density, the magnetic path will attain considerable uniformity and hence there will be a reduction in hysteresis. This is, of course, only one factor which will have this effect.

Mr. Richards queries the statement that the *LNF* alloy powders should give lower eddy-current loss than carbonyl-iron powder of the grade *E* type, but here again this has been proved by actual tests, particularly at very high frequencies. Also, the curve in Fig. 1 (*c*) of my paper, which is taken from a large number of production results, indicates that the carbonyl-iron powders have an advantage only at very low permeability values. The reason for this is the greater uniformity of particle shape and surface of the grade *E* carbonyl-iron powder, which is maintained in low-permeability cores because of the lower pressure used in manufacture. In the higher-permeability cores both the softer grade *C* carbonyl powder and the *LNF* powder are distorted in shape by the working-pressure values in order to attain the required permeability, and so the advantage of the original spherical shape of the carbonyl powder is partly lost.

Mr. Randall states that I have omitted Hametag, powder from Table II ; the reason is that this powder has not been used in Germany for making magnetic dust cores for over 20 years. This information was obtained from a tour I made of Germany last year in connection with this subject. Hametag iron powder was used in the first German powder core loading coils. The carbonyl-iron powders which were introduced in 1926 gave so much better results, chiefly on account of their lower hysteresis loss, and better yield of fine powders for reduction of eddy-current loss. As mentioned in other papers in the Symposium, Hametag iron is better for conventional powder-metallurgical applications.

The remaining points which refer to my paper have been adequately covered by the remarks of my colleague, Mr. A. D. Thomas, who kindly deputized for me at short notice as a result of my unavoidable absence when the paper was presented.

Mr. **S. J. Garvin** (Murex, Ltd., Rainham, Essex) replied : With reference to Mr. D. A. Oliver's question about the use of titanium hydride to assist in the manufacture of sintered alnico-type alloys, I should like to say that we have not employed this material in the conventional way or for the purpose for which it is famous in powder-metallurgy literature, that is, the evolution of nascent hydrogen during sintering to purify the powder compact and thus improve sintering conditions.

While a considerable amount has been said and written about this use of the hydrides, we at Murex have never had any real success with this technique. No marked improvement in the porosity of sintered compacts could ever be detected when one of the hydrides had been added to the powder mixture.

In view of the high cost of hydrides, and of titanium hydride in particular, the application of this technique was not further pursued. We have, however, used titanium hydride at times in experimental work when it was desired to add titanium metal to the alloy mixture. I understand that titanium hydride has been used in the manufacture of sintered alnico, but, instead of being mixed in with the alloy powder mixture and compacted, the titanium hydride was mixed with an iron–aluminium alloy. This powder mixture then took the place of what we have called the protective " getter " agents.

This is a perfectly logical application of the hydrides because they will certainly assist in the purification of the furnace atmosphere. As we achieved satisfactory results with a " getter " of Fe–Al, Ti–Al, or even aluminium powder alone, we felt that no particular advantage would be gained by making additions of metallic hydrides to our " getter " agent.

Dr. **L. B. Pfeil** wrote in reply : Mr. Randall comments on my remarks about the hysteresis of mechanically hard particles at low flux densities, and suggests that the reduced effect observed is due to the lower induction (presumably the effect of a lower intrinsic permeability). This may be certainly regarded as a further possible explanation ; it is felt, however, that one might also question the justification in assuming a lower intrinsic permeability ; in general our knowledge of properties at very low flux densities is rather scanty and there are other unknown factors to consider, such as the shape of the hysteresis loop. The main point that the author wished to illustrate was the fact that the conventional analogy of magnetic and mechanical hardness could not usefully be applied in considering materials used in dust cores. Mr. C. E. Richards referred to the low hysteresis of carbonyl-iron at relatively high flux densities ; it is believed, however, that in the application he probably has in mind, the flux density, although possibly a few hundred times greater than that normally used in radio applications, was far below values common in other applications, *e.g.*, conditions under which the region of magnetic saturation is approached.

Mr. Randall also referred, as did Dr. Chaston, to nickel–iron alloys produced by the carbonyl process and the two directions in which benefits might be obtained. Firstly, for dust cores, the production of a spherical alloy particle would undoubtedly be very useful ; attempts in the past to do this have not been very successful owing to the difficulty in obtaining homogeneous alloys at the comparatively low temperature used for decomposition. This problem is one to which more attention is being given. The development in Germany referred to by Mr. Randall involves the use of carbonyl powders as a convenient starting point for the production of nickel–iron alloys in other forms and of high purity and closely controlled composition. The important effect of the latter factors is well known, and such questions as cost and refinements of technique must decide whether such a process can offer advantages over more conventional methods.

Discussion on
Section D—HARD-METAL CARBIDES*

Dr. **C. Sykes,** F.R.S. (rapporteur), in presenting Section D of the Symposium, said : The subjects of the four papers with which I have to deal have been carefully selected, and there seems to be very little overlapping, so that it is possible to deal with each paper separately.

The first paper, by Mr. Brownlee, Dr. Geach, and Dr. Raine, deals with the preparation of carbides, and is concerned primarily with the carbides which are used commercially in this country, namely, tungsten carbide, titanium carbide, the mixed carbide, and also tantalum carbide. The technique described by the authors, as they point out, is the one which they use, and no doubt during the discussion it will be made clear that a substantial amount of these materials is made by other methods.

The production of tungsten carbide using the high-frequency furnace and a mixture of carbon, lamp black, and tungsten, calls for no special comment. The production of titanium carbide is a relatively more difficult matter. As the authors point out, the heat of formation of titanium oxide is very high, and therefore it is not an easy matter to produce a carbide which is free from oxygen. They carry out this procedure in relatively high vacuum, and operate at a temperature up to 1950° C., and in this way they produce a carbide which is substantially free from oxygen. This is a particularly difficult reaction, and the authors indicate that the preparation of the powder—that is, the titanium oxide and carbon —has to be carried out properly before the heat-treatment. They also describe a similar process for the production of tantalum carbide, and indicate that this procedure is a somewhat simpler one on a commercial scale than that for titanium carbide, owing to the fact that the heat of reaction in this case is somewhat smaller.

With regard to the mixed carbide, they describe again a vacuum process for treating suitable mixtures of titanium and tungsten carbide, and give data on the solubilities of tungsten carbide in titanium carbide and titanium carbide in tungsten carbide. The solubility of titanium carbide in tungsten carbide is small, whereas the solubility of tungsten carbide in titanium carbide is quite high. There is some discrepancy between the figures given for the solubility of tungsten carbide in titanium carbide in this paper and in the paper by Dr. Sandford and Dr. Trent. It seems possible that the vacuum method may enable the reaction to proceed more quickly between the solid powders.

Finally, the authors describe the preparation of the mixed powder, tungsten carbide, titanium carbide, and cobalt, and emphasize the importance of the correct milling of these powders. They indicate the amount of liquid which should be used and the ratio of the weight of the balls to the weight of the powder. This ball-milling operation is a very important one, and one which is not particularly well understood. I find it quite difficult to understand how the grain-size of a powder can be actually broken down in a ball mill, where in effect one must be dealing with a very sluggish motion of balls through what is in effect a very viscous fluid. Perhaps some further points in connection with ball milling will come out in the discussion.

Mr. Burden in his paper deals with the manipulation of the powder from the stage of the final mixing to the production of the final tip, and discusses the four operations concerned—pressing, presintering, shaping, and sintering. I think that the main points of interest in this paper are (*a*) the fact that the pressure applied to the compacts is relatively unimportant, and (*b*) the need for accuracy in the control of the grain-size of the powder in order to finish up with the correct dimensions. In most other powder-metallurgy operations, of course, it is also important to get a good control of dimensions, but it is particularly important in the case of these hard carbides, because the only method of adjusting the dimensions of the final sintered tip is, of course, by grinding. Mr. Burden also emphasizes the necessity for very careful control of the carbon–oxygen balance in the powder. That is also referred to by Dr. Sandford and Dr. Trent. This is, of course, the key technical

* The following corrections should be noted to the first edition of the Symposium :

Page 75, right-hand column, line 10 : The equation should read " $TiO_2 + 3C \rightleftharpoons TiC + 2CO$."

Page 76, left-hand column, line 11 : " The furnace is heated " should read " The furnace is evacuated."

Page 78, right-hand column, line 2 : " to oxidize the cobalt " should read " to reduce any oxidized cobalt."

Page 92 : The list of authors to the paper " The German Hard-Metal Industry," is incomplete and should include Mr. A. E. Oliver, of Hard Metal Tools, Ltd. Mr. B. E. Berry is associated with Murex, Ltd., and Mr. G. J. Trapp was with the Ministry of Supply when the investigation was made.

problem in the production of carbides, and is the reason why it is so important to produce the titanium carbides and the tungsten carbides free from oxygen in the first place.

Dr. Sandford and Dr. Trent deal with the physical properties of the carbides, and they describe in some detail the sequence of reactions which takes place in the pressed compact during sintering. An interesting point to which they refer is the very large grain-size of the cobalt powder. I think that the fact that the grain-size is of the order of 1 mm. was first brought out during the war, and the National Physical Laboratory did a good deal of work on that aspect of the matter. The authors also describe why it is that the shape of the tungsten carbide grains is so angular, and this is attributed to the precipitation of tungsten carbide from the cobalt matrix during the cooling operation. It would be interesting to have a similar description given for the structure of the tip containing mixed carbides.

The authors then describe experiments which they carried out in order to get a better idea of the rather complex equilibrium diagram associated with this class of alloy, and they conclude that if one deviates by more than 2½% from the theoretical carbon analysis, then, depending on which side of it you happen to be, one or other type of defect will be obtained. If there is too much carbon, graphite will tend to be deposited. The authors feel that this is not very important, except in very special cases, whereas Mr. Burden, who refers to the same thing, finds it objectionable. If the carbon is too low, the so-called η-phase appears, which gives a peculiar cored type of tip and leads to some brittleness in service.

The authors go on to outline a number of physical properties of the carbides, and show, for example, that the density is very near to the theoretical value, indicating that the porosity of these materials in production is very small indeed. They refer to "the inherent grain-size" as being very important in connection with the quality of the powders and the final tips, and this, I think, calls for some further elaboration. Two other points of interest are the very high compressive strengths of this type of material (from 250 to 400 tons/sq. in.) and the high modulus of elasticity.

The fourth paper deals with a visit made by some of the authors and other colleagues in the industry to Germany to examine the technique of the German carbide industry. In general, I think that the main conclusion of the authors is very similar to that of other metallurgical teams which have visited Germany, namely, that the Germans had a very large number of problems of a technical nature purely associated with supply, and these problems absorbed such a large proportion of their effort that they were not able to spend as much time as would otherwise have been the case on pure development work. The output of the German carbide industry was, of course, very large indeed, and they had to use inferior methods of producing the powders simply because there was not enough hydrogen to deal with the quantity of material required.

The team singled out, as being among points of interest, the reduction furnaces used by Krupps for the production of tungsten powder, a special vibratory ball mill which is claimed to have rather special properties, and a gadget developed by Krupps for the multiple hot-pressing of tips from substitute mixtures. Certain mixtures free from tungsten, consisting of molybdenum carbide, titanium carbide, and nickel, were used by the Germans as substitute materials, and the authors conclude that they were not very satisfactory in service.

It is interesting to hear that the Germans did a good deal of work on sintered alumina, but finally came to the conclusion that it was not really a satisfactory material. They also described the production of diamond wheels for grinding, in which diamond powder is impregnated in tungsten carbide and nickel.

On the general research aspect, I think that the main point of interest is the conclusion which they attribute to Dawihl in connection with the manufacture of the Widia Elmarid die, namely, that the η-phase, which is such a nuisance in ordinary tips, can be utilized if you go about it in the right way in the production of dies.

Towards the end of the paper there is a very abbreviated description of certain methods involving the use of electrical-conductivity and magnetic-saturation measurements, which Krupps claim give a good deal of information with regard to the properties of carbide tips. I think that it would be very useful to have that particular piece of information elaborated, with a view to seeing whether these methods can be applied to the examination of bodies of arbitrary shape ; in other words, can these electrical and magnetic methods be applied to tips, or must one have a nice round cylindrical bar ?

There are one or two notable omissions in the papers. Nothing is said of the production of tungsten powder itself, except for the German practice, and nothing is said of the use of the hot-pressing process in this country.

Mr. **R. Chadwick** (Imperial Chemical Industries, Ltd., Metals Division, Birmingham), opening the discussion, said : I had the privilege of joining one of the teams which investigated the German hard-carbide industry, and therefore I am very interested in the excellent summarized account, by Mr. Trapp and his colleagues, of wartime developments. The paper makes no mention of the use of tungsten carbide for projectile cores, although we were given to understand that during the war period this was of very considerable importance, a total of some 1500 tons of hard-carbide tips having been produced.

For projectile cores hot-pressing had very considerable advantages, for it was possible to press exactly to the finished size, except in respect of length, and so avoid the final grinding of an accurate radius on the tip of the core, which is essential in a product which has been cold-pressed and sintered. Hot-pressing technique had the further advantage that it was possible to use sintered scrap, either whole or coarsely broken, according to its size, which was, of course, impossible with cold-pressing techniques. Krupps made large quantities of tungsten

carbide tips of small calibre for experimental use at the time of the Spanish Civil War, and later these and other small calibres were discontinued in favour of considerably larger sizes of projectile. The scrap cores were successfully incorporated in larger projectile cores and Fig. A shows a section of a hot-pressed core of 22 mm. dia., incorporating a number of 6-mm. dia. cores. The whole product is homogeneous and of satisfactory density and strength, and the presence of the scrap material is revealed on an etched polished section only by careful adjustment of the lighting.

Peak production at a rate of some 100,000 per month of projectile cores was achieved in 1941–1942, when it comprised more than half of the Krupp and nearly half of the total German output of hard carbides. In 1943 the requirements of tungsten for machine tools had reached such proportions that it was no longer possible to spare tungsten for projectile tips, the manufacture of which was temporarily discontinued. Research on substitute carbides achieved success only in 1944, and no substantial production of tungsten-free material had commenced when the war ended. In view of this historical sequence it would therefore seem that the very extensive and complete Krupp plant for hot-pressing was in fact put down for projectile-core production, and not for substitute carbides.

Mr. **H. W. Pokorny** (Chemo-Metals (London), Ltd., London) : I should like to bring out one point made in the paper by Brownlee, Geach, and Raine, referring to the ball-milling process. I have made some studies with regard to this, and in view of the very inadequate knowledge of this aspect, I tried to produce a tungsten-carbide/cobalt mixture without going through the conventional procedure of ball-milling. The method adopted was as follows : Cobalt oxide of high purity, containing 76% of cobalt, was mechanically mixed with tungsten carbide. This mixture was reduced in hydrogen flow at a temperature of 850° C.

FIG. A—Section of hot-pressed core of 22 mm. dia., incorporating 6-mm. dia. cores (*see* R. Chadwick's contribution)

The powder was divided into three parts. No further processing was done with the first part. The second part was ball-milled for 4 hr. The third part was ball-milled for 48 hr. Test compacts were made from each of the parts at a pressure of 10 tons/sq. in. All pellets were sintered together in a batch furnace at a temperature of 1350° C. for ½ hr.

Examination of the sintered pellet showed uniform hardness for all, and the specific gravity to be the same within very narrow limits. Microscopical examination revealed coarse grain-size of the un-ball-milled powder, considerably smaller grain-size of the pellet made from the second part, and slightly smaller grain-size in the pellet made from the third part.

It has been found very difficult to give an explanation for this, but it is thought that it is not simply a question of mechanical disintegration of the carbide. There may be a certain amount of working hardness of the cobalt. It is thought, however, that the above experiments prove the old theory of ball milling not to have any foundation, as apart from the coarse grain-size, the pellets made from the unmilled mixture appeared to have normal structure and properties.

It would be very interesting to learn if anyone else has made experiments in this direction obtaining similar results, and can give an explanation for them.

Dr. **J. C. Chaston** (Messrs. Johnson, Matthey and Co., Ltd., London) : I find the excellent paper by Dr. Sandford and Dr. Trent very interesting, as going, probably, more thoroughly than has ever been done before into the mechanism of sintering carbide–cobalt compacts. There is one point, however, which I find it very difficult to understand. The authors mention that a final stage in sintering is required to ensure freedom from porosity. They point out that before this stage, central porosity exists as a series of microscopic holes filled with gas, and they say " Presumably this is eliminated by being taken into solution in the liquid phase, with subsequent diffusion to the surface." I wish that I could believe that such diffusion happened, because if it did it would provide an excellent method of degassing melts generally.

In a melt of cobalt, however, it is not general experience that hydrogen which is in solution will diffuse to the surface and escape ; and if it will not do it in a melt I do not see why it should do it in a sintered product. As an alternative suggestion of the effects of prolonged heating, I wonder whether some solution of the carbide in the cobalt occurs which alters the solubility of hydrogen in cobalt. I do not know whether that is possible, but it is a question that could be settled by experiment. It would be of interest, also, to determine definitely whether porosity in iron-copper compacts can be removed also by a process of prolonged heating. I do not think that it can.

Mr. **B. E. Berry** (Murex, Ltd., Rainham, Essex): I should like to compliment the various authors of the three papers dealing with methods of manufacture of carbide and the more technical research items, on the excellence of their papers, but I feel—and Dr. Sykes mentioned this point—that one particular application has been omitted, and that is the hot-pressing method. Almost no mention at all has been made of that, apart from a brief reference in the paper on German technique.

I think that the hot-pressing method is a very important method for the production of tungsten carbide hard-metal pieces, particularly as the tendency now is to produce bigger and bigger pieces of hard metal. The hot-pressing method thus becomes still more important. As applied to tungsten carbide the hot-pressing method is not, perhaps, precisely the same as is understood by the metallurgists who deal only with the ferrous metals or the non-ferrous, non-carbide metals, in that it can be carried out and is, in fact, always, I think, carried out in carbon moulds. The carbon mould is made to the shape of the finished piece and arranged so that all the shrinkage which occurs takes place in one direction only. It is possible, therefore, to hold dimensional tolerances in all directions except the direction of pressing, much more closely than is possible with the normal cold-pressing methods.

There are various hot-pressing methods. There is the method of heating by high-frequency current, and there is also the method of direct-heating by passing the current directly through the mould. I believe that American technique uses a combination of two methods, passing the current directly through a mould which at the same time has a molybdenum-wound furnace round the outside, so that the main heating is by direct current but there is subsidiary heating by the molybdenum-wound furnace round the outside of the mould, which will tend to even out the heating and to prevent heat losses. I am hoping that others in the industry may have a little more information about this American method of hot-pressing.

In the comparison of the high-frequency method with the direct-heating method, there are various points to consider. I think that with the high-frequency heating the two biggest difficulties which have to be overcome are these: First, there is the method of measuring the temperature. Apart from the method of measuring by sighting on the top of the mould with an optical pyrometer, which obviously is not very satisfactory, it is possible also to measure the temperature by means of a platinum–rhodium thermocouple let down into the mould in a hole drilled in the side of the mould. This again, however, is not altogether satisfactory, and it seems to me that probably the most reliable method of measuring temperature would be to have a hole through the side of the coil with a means of sighting either a radiation pyrometer or an optical pyrometer through the side of the coil on to the mould itself.

The set-up for the high-frequency method of heating is to place the carbon mould inside a fireclay sleeve and to pack round the mould inside the fireclay sleeve with an insulating material, generally a fine carbon powder. The high-frequency coil is then placed outside the fireclay sleeve. The method of determining the temperature obviously presents some difficulty.

The other point which has to be considered in the high-frequency method is the means of obtaining a really long even-temperature zone. Where it is required to press long, big cylinders of carbide we do need a very long temperature zone, and I think that a good deal of development in high-frequency-furnace design is required before it is possible to use that method for very long cylinders. The longest cylinder which we ourselves have pressed to date, using the high-frequency method, is about 7 in. in length and 3 in. in dia. Perhaps someone here can produce figures which are greater than those.

I have not any very great experience with the alternative resistance-heating method, and have only seen the set-up for that method in Germany, but there may be others with more experience who may be able to give us some idea of the advantages and disadvantages of that method. I think, however, that the question of hot-pressing is one which vitally concerns the carbide industry, and any expressions of opinion from those concerned on the relative merits of high-frequency heating compared with resistance heating, or the combination of resistance and molybdenum-wound-furnace heating, would be of very great interest.

I think also that expressions of opinion on the relative value of hot-pressing compared with the normal cold-pressing method, would be of interest where it is a question of dealing with pieces of a size which could conveniently be pressed by either method. My own feeling is that if there is no particular reason for wanting to use the hot-pressing method, then, of the two, I should prefer the normal cold-pressing, unless it is essential to keep the dimensions to very strict tolerances.

The temperature in hot-pressing is simply dependent on the sintering temperature of the particular mix which is being used; that is to say, a maximum temperature of about 1570° C. when dealing with normal grades containing titanium carbide. The sintering temperature for the hot-pressing method is the same for any grade as for the cold-pressing method. We are still very handicapped in this country in developing the hot-pressing method owing to the lack of really good carbon. There is no doubt that the carbon which is produced in this country, particularly in the bigger sizes, is not entirely suitable for use with the hot-pressing method.

In Mr. Burden's paper there is a brief reference to granulation. As we tend more and more to develop the method of pressing tips individually, rather than cutting from bar, and attempt to press tips and other pieces on automatic presses, the question of making the powder flow reasonably becomes very important, and I should very much appreciate any information which Mr. Burden or others can give on methods of granulating and the necessity for granulating. It may be that our experience differs from that of other people concerned, in that one of the factors which is

bound to influence the granulating is the method of waxing which is used. We use paraffin wax as the waxing medium ; others use petrol–camphor and various other materials. We have found by attempting to granulate by prepressing and screening when the granulated powder is used for pressing, a higher pressure must be applied than would be normal in order to be sure that the granules have been deformed and that the pressing is correct. I do not know whether that experience is common to others who have used this method. We try as far as we can to avoid having to granulate. Frequently we take a fairly heavily waxed powder, screen out certain fractions of that powder, and blend them to give the pressing properties required. Any information on this subject would be of interest.

Mr. **H. N. Blyth** (Messrs. L. M. Van Moppes & Sons, London) : We have had to use hot-pressing in the process with which I am connected, and we are hampered all the time by the fact that it has to be done in carbon, and that carbon will not stand heavy pressures. If anyone could give information about an alternative material or a better method of bonding the carbon I should be interested, and I think that it would be of interest to all users of hot-pressing. I turned a small mould out of a piece of zirconia refractory brick and made a pressing in it. It stood a pressure of about 6000 lb., and temperature made no difference to it, but the surface was extremely poor, and it was no good as a substitute for carbon. If there is any information on a better form of bonding of carbon or on a substitute for carbon, I should be very interested to hear it.

Mr. **A. G. Metcalfe** (Queens' College, Cambridge) : I should like to congratulate the authors of these four papers on bringing together in such a readable form all the available information on the production of cemented carbides. Much of it, I think, has never been published before.

There are two points which I should like to raise. The first relates to a discrepancy to which Dr. Sykes called attention, and is between the results given for the solubility of tungsten carbide in titanium carbide, and the second point concerns the interesting results found by Brownlee, Geach, and Raine in the TiC–TaC system. From its resemblance to the copper–silver system it would appear to be possible to obtain precipitation hardening at the tantalum carbide end of the TiC–TaC system, and some precipitation on a fine scale seems to have been observed by these authors. In the case of cemented carbides, hardening might be obtained in either the bond (*see* Fig. 2, page 85) or the carbide grains ; and, whilst it is possible to prevent the carbides dissolved in the cobalt from being deposited on existing carbide grains by very rapid cooling, this method is not practicable owing to the attendant cracking. Hardening of the carbide grains would probably be a more suitable and effective process, since the rate of cooling would be expected to be slower than that required to retain the solid solution of the carbides in the cobalt. It is possible that it would not be so effective as hardening of the bond, particularly in the higher cobalt alloys. Actually I believe that no precipitation hardening has been observed in carbides, but it seems to be a very interesting process, which could lead to developments in that field, and it would be of considerable interest to know whether any such investigation of the titanium-carbide/tantalum-carbide system has been made by the authors from this point of view.

The second point concerns the solubility of tungsten carbide in titanium carbide, the results for which show some discrepancy as between the paper by Brownlee, Geach, and Raine, and that by Sandford and Trent. In the first of these papers the results are given on page 77, and in the second paper they are given on page 90. Many investigations have now been made of the so-called maximum solubility of tungsten carbide in titanium carbide, which appear to agree very well. To quote some of these results, the Russian workers, Molkov and Zwikker, obtained 82·5% ; Zumbusch and Sanders give 80%, whilst Brownlee, Geach, and Raine give 80–82%. In some cases, however, no temperature is given. On the other hand, Umanskii and Khidekal* give the following results for the variation of the solubility with temperature, which agree very well with the figures given in the paper by Sandford and Trent :

Temperature, °C	Umanskii and Khidekal	Sandford and Trent	Brownlee, Geach, and Raine
2000	81	82	75
1800	78	...	71
1600	75·5	75	61
1500	74·5	...	42·5
1400	74	74	...

Actually, the agreement between the first two sets of results is misleading, because at higher temperatures there is considerable disagreement, but so far as that is concerned the results have not yet been published. It can be seen, however, that these sets of results are not in agreement with those quoted by Brownlee, Geach, and Raine. It might be mentioned here that a solubility of 74% at a temperature of 1500° C. has been obtained using electrical-resistance measurements, confirming the results quoted by Sandford and Trent, which were obtained by the X-ray diffraction method. The results given by Umanskii and Khidekal, and by Sandford and Trent in their paper, were obtained by approaching equilibrium from the side rich in tungsten carbide. This would seem to be a more accurate method than heating loosely tamped powders of the two carbides.

In spite of these remarks, I am not clear whether Brownlee, Geach, and Raine refer to equilibrium or to the solubility obtained after the given times of sintering at the stated temperatures, and I should be grateful for any information which they can give me to clear up that point. They refer to " effective solubility," and I am not clear what that means.

Finally, are any of the authors able to give any information on the existence of the various compounds which have been claimed to exist, such as $Ti_2W_2C_4$ or Ti_2WC_4 ?

Dr. **E. A. Bano** (Electro-Alloys, Ltd., London) :

* *Zhurnal Fizicheskoy Khimii*, 1941, vol. 15, p. 997 ; *Journal of Physical Chemistry*, 1942, p. 980.

The hot-pressing process leads to the study of equilibrium diagrams at pressures higher than atmospheric. I should like to know whether this problem has been given sufficient thought, and what scientific and technical consequences it would entail.

Mr. **D. J. Jones** (G.E.C. Research Laboratories, London) : Brownlee, Geach, and Raine refer to the use of absolutely dry hydrogen. This is very important in powder metallurgy in general, but even on a laboratory scale it is extremely difficult to maintain consistently a very low water-vapour content. In this connection I should like some information on the method employed for determining the low water-vapour content required.

With reference to ball milling, the authors prefer wet to dry milling, claiming that better mixing is obtained. It is difficult to ensure thorough mixing of powders, particularly with powders of different densities, as, for example, tungsten and molybdenum. In the case of these two powders, a comparison of wet and dry milling has been made on a number of occasions, but I have been unable to arrive at a definite conclusion as to which is the better method. It would be useful to have the views of other members present on the respective merits of wet and dry milling.

The other point I want to make refers to the use of stainless-steel mills. It is interesting to note that the amount of iron pick-up is surprisingly small. Is there any chromium pick-up ? Also, how does the wear of this mill compare with a tungsten carbide mill ?

Referring to the paper by Mr. Burden, there are two minor points I should like to raise. Firstly, under the heading " Pressing," he refers to an average grain-size of 8 μ, and also to a grain-size of 0·2 μ. Should the last figure be 2·0 μ or is it actually 0·2 μ ?

The second point refers to the effect of pressure on the resulting density when sintering occurs in the presence of a liquid phase. A characteristic feature of this type of alloy is that pressure has very little effect on the resulting density. In fact, almost theoretical density can be attained with uncompacted powder. In view of this, why quote a minimum pressure of 0·25 tons/sq. in. instead of making this figure zero ?

Mr. **Josiah W. Jones** (College of Aeronautics, Cranfield, Bletchley, Bucks.) : One of the papers presented yesterday afternoon dealt with engineering applications, and I wondered, reading through it, just what the engineering interest was. I had not then visited the exhibition which is being held in another room, which more than justifies the title of the paper. The attractions at the moment are very limited. There appear to be improved properties as the result of the sintered and compressed structures in contrast to metallurgical structures—improved wearing of bearings and improved electrical properties, and, of course, improved hard-cutting properties.

That is by way of introduction to what I really want to say as to the hints contained in the paper by Sandford and Trent on engineering properties in which all engineers, and particularly aircraft engineers, are very interested. There is first the question of the improved Young's modulus (page 39). This is a very ancient practical problem, and one which is very difficult to assess and on which it is very difficult to obtain data. It is all a matter of very long-time experimental work, and theoretically, by our ordinary means of alloying small quantities of B into A, there does not seem to be much theoretical prospect of improving Young's modulus.

We are interested in the Young's modulus as distinct from the proof stress to give improved stiffness to a structure and to increase resistance to buckling stresses. Unfortunately, we are mainly interested in sheet, and in thin sheet at that, *e.g.*, 0·023 in., so that it seems rather looking into the future to anticipate stressed-skin construction from the materials offered here. However, I throw out the hint to this Institute.

Then, of course, there is the demand for improved creep properties. Nothing is mentioned in this paper about creep, and the absence of it, in view of its importance, leads one to suspect that there is considerable difficulty somewhere. However, the authors do quote remarkably high compressive strengths, and the fact that the transverse strength is mentioned gives rise to faint hopes that there may be some possibilities also in tension. I should like to know

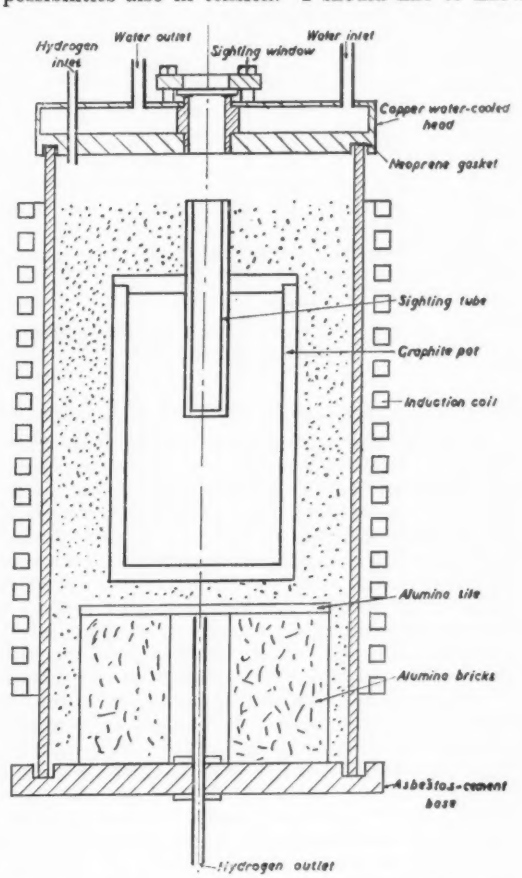


FIG. B—High-frequency-furnace coil set-up, using top-sighting window

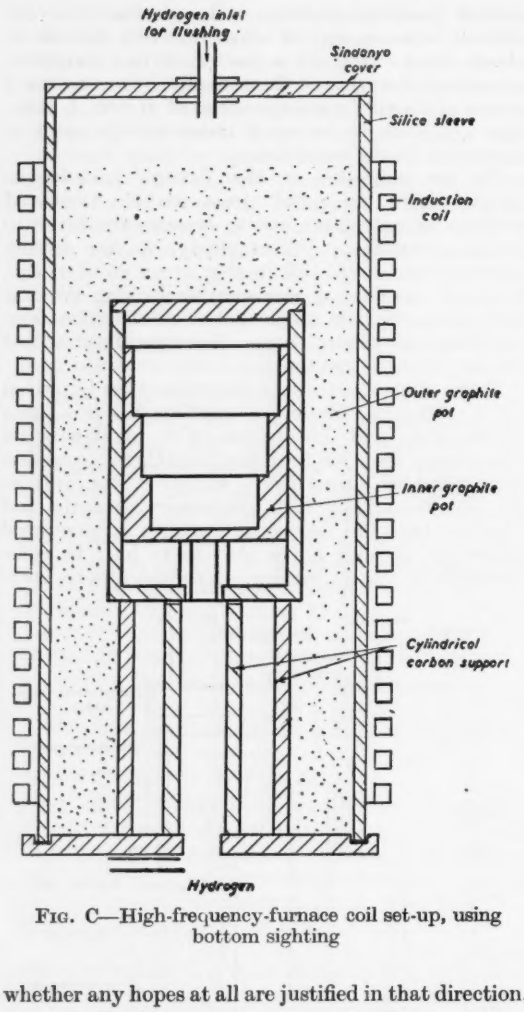


FIG. C—High-frequency-furnace coil set-up, using bottom sighting

whether any hopes at all are justified in that direction.

Mr. **D. H. Shute** (Production Tool Alloy Co., Ltd., Bedford) wrote: We were very interested to read the details of high-frequency technique used by Mr. Brownlee in the preparation of carbides. This is an aspect of high-frequency heating which has apparently received but scant detailed attention in the literature so far. This is somewhat surprising in view of the wide range of uses for high-frequency heating in carbide manufacture, from the preparation of the powders, sintering of cold-pressed compacts, hot-pressing, to the brazing of sintered tips on to shanks.

A note on our own technique may be of interest. The set-up shown in Fig. B. has proved very satisfactory when used in conjunction with a motor-generator high-frequency set. A silica sleeve fits into the furnace coil, both ends of the silica sleeve being ground flat. The top fits into an annulus in the copper water-cooled head, and a neoprene gasket makes a gas-tight seal. The bottom fits into an annulus in a 1-in. thick Sindanyo asbestos-cement disc. Sealing at the bottom is made by ordinary pipe jointing compound.

The graphite crucible rests on alumina bricks with a layer of insulating powder between, the thickness depending on the temperature required. Alumina powder is used for temperatures up to 1500° C., carbon powder being used for any higher temperatures. The sighting tube in the crucible is lined up with the sighting window. Hydrogen is fed through the water-cooled head and escapes through the tube in the bottom. This method is more economical of gas than the top-vent tube shown in Fig. 2 on page 74 of the Symposium.

When using a valve high-frequency set it is undesirable to have a large mass of metal near the work, and the simple set-up in Fig. C has been found satisfactory for sintering. Sighting is from the bottom of the crucible so that no sighting window is necessary. A disc of Sindanyo cement board seals the top of the sleeve. The joint can be made gas-tight with ordinary pipe jointing compound.

The sleeve is filled with hydrogen through the tube in the top. When the tube is completely filled with hydrogen this tube is sealed and the gas is then fed from the bottom. Provided that the sighting hole is kept small there is no trouble in preventing the entrance of air. Also, it is not necessary to provide a means of holding down the lid because there is no pressure of gas inside the sleeve.

This method of sighting from the bottom of a pot can be readily applied to hot-pressing, in conjunction with a down-stroking press.

The inner stepped carbon pot is raised from the bottom of the outer pot in order to eliminate temperature gradients in the charge and to approximate black-body conditions on the sighting area under the inner pot.

Sighting from the top is generally preferred for processing powders, and bottom sighting is better for sintering.

On page 75 of the Symposium, in considering the impurities present in titanium dioxide used for preparing titanium carbide, no mention is made of the presence of alkali salts. Commercial grades of TiO_2 made in Great Britain usually contain up to 0·3% of alkalies. In our experience the presence of a small amount of alkali is of considerable assistance in accelerating the reduction of TiO_2.

In the discussion the question of determining the moisture content of hydrogen was raised. Our own method of determining moisture, whilst not giving an absolute value of water content, has proved useful for routine control and may be of interest. The method consists simply of passing a measured quantity of hydrogen through a weighed glass-stoppered **U**-tube containing magnesium perchlorate. With a flow of 20 litres/hr. it is not necessary to use more than one **U**-tube, since at least 98% of the water absorbed by the magnesium perchlorate is picked up in the first tube. The hydrogen is metered through a sensitive sapphire-orifice flow-meter.

Our results over a number of years have shown that the claims made for one well-known type of drier are not substantiated in practice.

In considering the variation of the solubility of WC in TiC which was given by Sandford and Trent as increasing from 75% at 1600° C., to 82% at 2000° C., and

by Brownlee, Geach, and Raine as from 61% to 75% for the same temperatures, a useful comparison can be made with our own material which is representative of commercial practice at 1800° C. The analysis of this powder shows a combined-carbon figure of 8·0%, titanium 16·8%, and free carbon 0·35%, and X-ray analysis shows only the cubic lattice of TiC, the lattice parameter being 4·261 kX.

It is in our opinion a matter for conjecture whether the TiC or WC is satisfied with carbon; the fact that there is no trace of the W_2C lattice is inconclusive since we have no evidence of the solubility of W_2C in TiC. This fact makes it difficult to estimate the exact TiC content of the powder and may give rise to apparent discrepancies in the limit of solubility.

The back-reflection X-ray photograph in Fig. 6, facing page 86 of the Symposium, showing the large grain-size of cobalt, is also unusual in the grain-size of the WC. Back-reflection photographs of similar materials that we have taken show a larger grain-size of the untouched skin of a sintered piece, but a much finer grain throughout the bulk of the sample, an etched finish being necessary to show the cobalt at all on exposures made on other than the untouched surface.

AUTHORS' REPLIES

Mr. **H. Burden** (Brown-Firth Research Laboratories, Sheffield), in reply, said : Mr. Berry has asked for more information on the granulating of carbide powders for the pressing operation. We have found that owing to their fine grain-size carbide powders could be granulated very readily. The process consisted of making a cake or compacted mass of the powder, which is broken up by forcing the material through sieves of the desired size. The resultant product is then fed into a rotating container, which may conveniently be a ball mill, without a charge of balls, for a period of, say, 30 min. The latter operation improves the uniformity of size of granules and the flow characteristics of the granulated material.

The cake or compacted mass referred to above may be made in two ways. Firstly, the powder, the desired percentage of wax, and a solvent are mixed to form a mud ; the solvent is then driven off by heat, leaving a friable cake. A second method consists of preparing the waxed powder in the normal way and pressing this powder in suitable large dies at a pressure of ½–6 tons/sq. in. to give pressings which can then be broken up as already described.

The hardness of the granules, *i.e.*, the difficulty with which they break up in subsequent pressing operations, depends upon the method employed. The softest granules result from making a cake using the solvent-and-wax method, while harder granules may be obtained by pressing the waxed powder in a die. The hardness in this latter case can be controlled by varying the pressure applied.

It is true, of course, that granulated powders require greater pressures to form satisfactory green pressings than normal powders, but the increase in pressure need not be high. In any case, it is easy to check from the fracture of a green pressing whether granulated powder has compacted satisfactorily.

Mr. D. J. Jones questioned the grain-size figures mentioned in my paper, and in particular the figure of 0·2 μ given for the grain-size of tungsten powder. The figure of 0·2 μ is correct, and while it is difficult to say how accurate is the measurement of the grain-size, it is certainly of that order. In the exhibition of powder-metallurgy products there was a photograph of the tungsten powder in question taken by an electron microscope at a magnification of 10,000 diameters and an inspection of that photograph would confirm our estimate of the grain-size.

Mr. Jones also raised the question of the compacting pressures, and asked why one should give a range commencing at 0·25 tons/sq. in., rather than state that even uncompacted powder will sinter satisfactorily. This last statement is not quite true ; there is some limit to the voids which can be closed by sintering, and in a loose heap of powder there may be voids of such a size which would not be satisfactorily closed. A very slight pressure is necessary to close them, and a pressure of 0·25 tons/sq. in. is about the lowest limit which can safely be used.

I should like to take the opportunity of discussing one or two points from the other papers on hard-metal carbides. In the first place, I think that Brownlee, Geach, and Raine raise a rather controversial point in stating that the carburizing of tungsten on a large scale can best be carried out in a high-frequency furnace. In carburizing large batches of tungsten in a high-frequency furnace there is a serious danger of overheating the outside of the charge before the centre of the charge is properly carburized. This is due to the fact that a mixture of tungsten and carbon has poor electrical and heat conductivities. While it is possible to arrange for the heat-treatment cycle to cause effective carburization of the material from the outside to the centre of the charge, in so doing the outside of the charge, being at high temperature for a longer time, is subject to grain growth. In other words, if a large charge is used there is likely to be considerable lack of uniformity in the material with respect to grain-size.

On the other hand, when carburizing tungsten carbide in molybdenum-wound or carbon resistance tube furnaces of relatively small diameter, this effect can be held to a minimum, and owing to the fact that it is possible to arrange for material to be continuously fed through furnaces of this type, large outputs may be obtained in spite of the small diameter. I would suggest therefore that to obtain large outputs of tungsten carbide of the highest and most uniform quality the carbon tube or molybdenum-wound tube furnace is the preferred equipment.

I should also like to mention the difference in solubility of tungsten carbide in titanium carbide reported by the various workers. Brownlee, Geach, and Raine have mentioned that some further details should be given of the methods used in determining these figures. I think that is particularly desirable ; we find that many factors affect the amount of tungsten carbide which goes into solution in titanium carbide. One point on which it is necessary to disagree with Sandford and Trent is their statement that solution occurs readily. Under certain conditions a saturated solution of tungsten carbide in titanium

carbide can be formed, but in normal production processes it is very difficult to ensure that the maximum amount of tungsten carbide is in solution at any given temperature. It seems to be affected, as suggested by Brownlee, Geach, and Raine, by grain-size, and also it seems to be affected very much by the presence of small amounts of impurities—traces of iron, nickel, or cobalt cause much more rapid solution, and in production processes a small percentage of these metals is sometimes added to enable the solution to proceed more rapidly.

It is significant that no mention has been made in any of the work in this country of any compound formed between tungsten carbide and titanium carbide. That is the experience of our own work; and it seems highly improbable that any compound exists.

One other feature which has not been made clear in the papers so far is that while tungsten carbide shows increasing solubility in titanium carbide as the temperature is raised, no mention is made as to whether the reverse is true, *i.e.*, whether tungsten carbide is precipitated as the mixed crystals are cooled. In our own work we have found that this precipitation is very sluggish at temperatures normally employed in the manufacture of carbides, and it is considered that more work will have to be carried out on the tungsten–carbide/titanium–carbide system before the tungsten–titanium carbide hard-metals are as thoroughly understood as the tungsten carbide hard-metals.

Dr. **E. M. Trent** (Hard Metal Tools, Ltd., Coventry), in reply, said : Our paper describes, so far as we know them, the metallurgical processes which go on during the sintering of tungsten carbide and cobalt alloys. We did not deal in detail with the sintering of alloys containing titanium carbide. This is a very complex subject which is not so thoroughly known, and also it involves a more complex system, requiring much more space than we had available to deal with it. I hope, however, that this work will be done, as it presents certain very interesting metallurgical problems.

The inherent grain-size is introduced as a useful conception, and the ideas behind it will have to be pursued further. Here I think that we should be explicit as to exactly what we mean by grain-size. When we refer to grain-size we refer to the grain-size or particle size of the carbides, and not to the grain-size of the matrix or cobalt metal. By referring to inherent grain-size we mean that, while it is possible to alter the grain-size during the sintering process and to get grain growth of the carbide during that process, it is not possible to get a fine-grained carbide in the final sintered product unless an inherently fine-grained carbide is used initially. To maintain consistent grain-size entails the control of a very large number of factors.

Some question has arisen of the importance of graphite precipitation in those alloys which have an excess of carbon over and above that required to produce the compound WC. If the amount of graphite is small, we consider that it is not very important and it does not have a big effect on the mechanical properties of the material unless the amount of graphite precipitated is excessive.

Another point is that examination of the microstructure of these materials gives an exaggerated idea of the amount of graphite present. What is actually only an extremely minute amount will, on first polishing, appear to be a large amount because the graphite partly surrounds the carbide grains which fall out during the polishing. This is the same phenomenon that has been observed with graphite in cast iron, only it is more pronounced with these carbides. To get a true idea of the amount of graphite present, prolonged and careful polishing is necessary.

With regard to the elimination of final porosity, it is quite likely that the hydrogen solubility in cobalt is considerably affected by tungsten carbide present in solid solution. I can see no other way in which the small amount of porosity at the centre of sintered pieces can be eliminated in the final stages of sintering than by solution of the gas—which may be carbon monoxide, not hydrogen—in the molten metal and diffusion to the surface.

The question of the engineering properties of cemented carbides is a very interesting one. These metallic carbides have very interesting engineering properties and superlative ones in some directions, *e.g.*, very high Young's modulus, extremely high compressive strength, etc. These properties are partly the properties of the carbides themselves, and it is essential that in the future more fundamental work on the structures of the metallic carbides—their atomic and electronic structures—should be carried out. It is partly the properties of the carbides and partly the properties peculiar to these alloys, in which, as we have tried to illustrate in our paper, the carbides and the softer binding metal both form continuous interwoven networks in three dimensions, which produce a particularly rigid type of structure. There is a great deal of fundamental work on these materials which should be carried out because of their very unusual engineering properties, but at the same time they have certain poor properties as well—in particular the lack of ductility. Perhaps further research will improve, if not the ductility, at least the shock resistance, of these carbides still further.

Questions have been raised on the other papers with regard to the preparation of the powders by ball-milling and with regard to stainless-steel ball mills. We find that the pick-up of iron from stainless-steel mills is very small, and have had stainless-steel ball mills which have had a life of over 10 years.

The paper by Brownlee, Geach, and Raine, on the preparation of carbides, is an excellent one, and I should particularly like to refer to the pioneer work which they have carried out in this country on the preparation and sintering of these materials in vacuum at very high temperatures. This vacuum technique which they have developed seems to me to have very important possibilities for the powder-metallurgy industry, and I feel that the fact that it has to be carried out in batch processes is not necessarily a drawback.

On the question of milling, I think that it is partly a question of the breakdown of grain-size, but probably only to a minor extent; it is much more a

question of the break-up of aggregates. There has been a suggestion that the carbide particles become coated with cobalt, but personally I cannot see how that could happen, and I have not seen any concrete evidence for it.

One of the most important things that comes out of Mr. Burden's paper is the necessity for maintaining consistency in the practice adopted for the production of these carbides. That is very important if consistent properties in the final product are to be obtained.

With regard to the paper on the German hard-metal industry, and the discussion on page 97 of the Widia Elmarid composition, it should be pointed out that the phase which in Germany is known as the δ phase is the one which in this country we refer to as the η phase.

Mr. **B. E. Berry** (Murex, Ltd., Rainham, Essex), in reply, said : In preparing a paper of this kind it is obvious that we had to select a few points which we considered to be of particular interest and to confine these within reasonable limits ; therefore we purposely omitted more than a bare reference to the manufacture of projectile cores by the hot-pressing method, which Mr. Chadwick mentioned. Details of the German practice are to be found in the B.I.O.S. report* of our investigation, which is just about to be published, and I do not think that I need add very much at this stage, except to say, with reference to Mr. Chadwick's Fig. A (page 552), that we saw similar examples of the hot-pressing of smaller cores into bigger ones.

On the general question of the relative merits of hot-pressing as compared with cold-pressing for the production of these cores, I still feel that the cold-pressing method is better, though I know that the other method was almost exclusively used in Germany and was used to some extent in this country. We ourselves used it to a small extent for this purpose, but I think that the advantages generally are in favour of the cold-pressing method.

Dr. **T. Raine** (Metropolitan-Vickers Electrical Co., Ltd., Manchester), in reply, said : One of the points raised in the discussion was that in our paper we should have dealt with the preparation of tungsten powder. That, however, is a complete subject in itself and, as these papers are limited to about 4000 words we thought it best to leave that out. I should say, however, that, as mentioned in the paper by Dr. Sandford and Dr. Trent, the grain-size and grain shape of the tungsten powder used have a very important effect on the resultant properties of the carbide.

Dr. Sykes referred to the carbon/oxygen reaction during sintering as being extremely important. We entirely agree with that, and we find that by a combination of vacuum and hydrogen sintering, which is very simple to carry out, it is possible to control the rate of carburization during sintering without any difficulty.

The chief point which has been raised so far as our paper is concerned is Mr. Metcalfe's point about the solubility of tungsten carbide in titanium carbide. In that connection, I think that none of the authors who have dealt with this subject have published their experimental data. If these were published it might be possible to throw more light on the matter. In our case, we took the separate powders, mixed them together, and heated them, first to 1500° C., then to 1600° C., and then to 1800° C., and so on, determining the solubility limit at each stage by X-ray methods. I should point out that the figures to which Mr. Metcalfe has referred relate to actual production batches weighing about 50 kg., but we have found that by working with very small batches and compressing them before heating, we can obtain a solubility of 80–82% of tungsten carbide in titanium carbide, and this fact is mentioned in our paper. We understand that Mr. Metcalfe used a different method to make the solid solutions. He first heated selected compositions to a very high temperature, so that they were single phase, and then gave these alloys prolonged soaking at lower temperatures, causing precipitation of tungsten carbide to occur, and then measured the solubility limit.

We approached the problem from the other extreme, that is, we allowed the mixed carbides to diffuse into each other by prolonged heating at a given temperature.

Obviously, the two methods will give the same results only if complete equilibrium is reached. At temperatures so far below the liquidus, the diffusion and precipitation rates are so slow that abnormally long annealing times will be necessary to obtain complete equilibrium. We chose the method of diffusion, rather than precipitation, to explore this system, since this is the mechanism used in the manufacture of sintered carbides.

Some reference should also be made to the grain-size of the powders used, since this will have a considerable effect on the rate of diffusion of the carbides, a coarse grain-size slowing down the diffusion and leading to lower values for the solubility of tungsten carbide.

Reference has been made to the use of pure dry hydrogen. We find it very difficult to obtain dry hydrogen on a large scale, and therefore we use the vacuum process.

As regards wet ball-milling *versus* dry ball-milling, I think it is agreed in the industry that wet milling gives a better product, and a product which is more free from porosity than dry ball-milling. With regard to stainless-steel ball mills, we have used both stainless-steel ball mills and carbide-lined ball mills. The life of the carbide-lined mills is several years. The stainless-steel ball mills which we have had in operation for about nine months, have still considerable life left in them.

Concerning the pick-up of iron, according to our own analysis the iron pick-up after seven days' milling is of the order of 0·1–0·2% of iron.

Dr. **E. J. Sandford** and Dr. **E. M. Trent** (Hard Metal Tools, Ltd., Coventry) wrote in reply : With regard to Dr. Sykes' presentation of Section D of the Symposium, we are glad to confirm that the large grain-size of the cobalt-rich phase in sintered carbides was first pointed out in a report issued by the National Physical Laboratory. We have since confirmed that this is so. We find that X-ray back-reflection films

* B.I.O.S. Report No. 1385.

taken on fractured, polished or ground surfaces often show no spots from the cobalt-rich phase and it is usually necessary to remove some of the carbide by electrolytic etching with sodium hydroxide solution.

As regards the limits within which the carbon content must be controlled, we wish to point out that the figure of $2\frac{1}{2}$ refers to a percentage of the total amount of carbon present. Thus an alloy with 6% of cobalt would have a theoretical carbon content of 5·75%, and the greatest allowable deviation would be $2\frac{1}{2}$% of this, namely, down to 5·60%.

The idea of " inherent grain-size " was introduced to indicate a phenomenon which deserves further study. Pieces sintered from some batches of powder show marked grain growth of the carbide, which becomes more serious as the sintering temperature is raised, whereas other batches of the same composition show much less grain growth at the same temperatures. The explanation probably lies in the grain-size distribution of the powder, and in fact it has been found that a reduction in the powder grain-size beyond a certain point apparently causes an increase in the grain-size of the sintered piece. The surface condition of the powder may also be a factor, but little or nothing is known of this.

We agree with Mr. Berry that the lack of good carbon for hot-pressing moulds is a serious handicap to the hard-metal industry of this country, and we hope that the carbon manufacturers will devote attention to this in the near future. The question of dimensional accuracy raised by Mr. Chadwick depends of course on the surface finish of the graphite mould. We also agree that the cold-pressing technique is generally easier to apply when the size and shape of the piece permit, but hot-pressing possibly gives better properties when very impure raw materials are to be used, and of course it is almost essential when very low binder contents are used. Since the German projectile cores usually contained only 3% of nickel, the use of the hot-pressing technique would be preferable. The comparative properties given by the two techniques are, of course, important. For small sizes, grain growth is generally less in hot-pressed pieces, but with large sizes this advantage is lost because of the longer heating time required. With high-cobalt alloys some of the cobalt-rich phase is squeezed out, and with a poor technique this may seriously alter the composition of the alloy. Even if the powder used in hot-pressing is of normal carbon content, the core of the pressing may contain η-phase. Since hot-pressing is carried out without a protective atmosphere, the initial stages of heating are accompanied by some oxidation of the powder, which is subsequently reduced by the carbon of the carbide. In the last stage of sintering, the surface is recarburized by contact of the molten phase with the walls of the mould. Thus changes in composition during hot-pressing generally include a lowering of the cobalt and carbon contents.

In reply to Dr. Chaston's remarks concerning the elimination of porosity during sintering, we can see no other explanation for this than that which we have given. It is not necessary, however, for the gas which fills the pores to be hydrogen. It seems probable that it contains carbon monoxide and possibly hydrocarbons. If the partial pressure of these gases in the surrounding atmosphere is low, they will be gradually lost from the exposed surfaces of the pieces. It is, of course, probable that the solubility of gases in the molten phase containing tungsten and carbon is considerably different from that in pure cobalt. The solidification temperature of this alloy is more than 160° C. below that of pure cobalt.

The results of X-ray work which we gave at the end of our paper were largely the work of Mr. A. G. Metcalfe, carried out in the laboratories of Hard Metal Tools, Ltd., and a fuller account of this work has been published recently.*

In reply to Dr. Bano, we should expect from phase-rule and other considerations that the pressures normally used in hot-pressing (up to 1 ton/sq. in.) would have no appreciable influence on equilibrium between metals.

The remarks of Mr. J. W. Jones on the requirements of engineers, particularly those of the aircraft industry, for new materials are of considerable interest. We feel that the unusual properties of cemented carbides, such as the high modulus of elasticity, will find applications. For example, the accuracy obtainable in boring sometimes depends on the elastic bend of the boring bar which may have to be quite long ; the use of solid-carbide boring bars reduces the bend to less than half for an equal tool load and is therefore an advantage. The difficulties of fabrication in the shapes and sizes required and the lack of appreciable ductility are two of the drawbacks to a wider application of sintered carbide, and it would appear at present that the production of thin sheet will not be possible for a long time. The creep properties of carbides are being determined but it is too early to give any results.

In reply to the correspondence from Mr. Shute, the lattice parameter quoted by him is much lower than that obtained by Metcalfe (*loc. cit.*) who gives a parameter of 4·297 kX. for the saturated solid solution at 1800° C. From the figure given, it would appear that the carbon content of Mr. Shute's carbide is about 1% lower than the theoretical (9·1%) for his composition. It may be that a high oxygen content has reduced the lattice parameter, and between 1 and $1\frac{1}{2}$% of oxygen by weight could account for such a low parameter.

From Mr. Shute's remarks on the back-reflection X-ray photograph reproduced in our paper, it is not clear whether he is referring to the grain-size of the tungsten carbide or of the cobalt-rich phase. In general, it would not be unexpected to find a larger carbide grain-size on the extreme skin of a sintered piece than in the core.

* *Journal of the Institute of Metals*, 1947, vol. 73, June p. 591.

A further reply by Mr. G. J. Trapp, which was received just before going to press, will be found on page 578

Discussion on Section E—POROUS METAL COMPONENTS*

Dr. **W. D. Jones** (rapporteur), in presenting Section E of the Symposium, said : One of the peculiarities of the technique of powder metallurgy is the ability to make products which are porous. Whilst on many occasions porosity in metals is not regarded as a good thing, there are occasions when it can be turned to advantage. Of course, cast materials are also sometimes porous, but in the powder-metallurgy article we do know where that porosity is and how big it is, which is not always the case with cast materials.

The three short papers in this section of the Symposium deal with porous products. Very little has been published on the structure of porous bronze bearings, and Mr. Carter and Mr. Metcalfe add to our meagre knowledge on this point. They have examined three typical porous bronze bearings, designated types *A*, *B*, *C*, under the microscope and also by back-reflection X-ray diffraction. Types *A* and *B* contain α solid solution, and the δ-phase $Cu_{31}Sn_8$. Type *C*, however, contains only α solid solution. The tin content of the three samples varied from 8·57 to 12·36%, and there was from 0·95 to 1·66% of graphite, together with small amounts of iron, lead, and zinc.

These authors made experiments by sintering compacts of 90% electrolytic-copper and 10% atomized-tin powders mixed together, with and without 1% of the copper replaced by graphite, and using either coarse or fine tin powders. The speed of diffusion of the two metals in these compacts is, of course, greatly increased by the fact that the sintering temperature is well above the melting point of tin. The authors point out that the compacts containing the fine tin powder reached structural equilibrium more gradually above 800° C. than those containing the coarse powder, which seems to me to be rather strange. That point is illustrated in Figs. 4 and 5 of their paper. The sintering time has little effect on the results, but it is noticeable that the graphite delays the attainment of equilibrium. The sintering temperature is much more important, particularly with reference to the peritectic temperature at 798° C. Below this temperature the γ and β phases exist, and appear in the final product as α plus δ eutectoid. Above this temperature only α exists in equilibrium with the liquid. Below this temperature the final product consists of interconnecting porosity, and above it the pores are isolated.

The exact nature of the porosity in porous bronze bearings is presumably a matter of considerable importance, and one would expect that the more satisfactory bearing would be made with an interconnecting porosity. The compressive strength of the bearings reaches a maximum as the sintering temperature rises to and above 800° C.

The other two papers in this part of the Symposium describe two different applications of porous powder-metallurgy products. The paper by Mr. Sinclair describes the preparation and properties of porous bronze filter materials containing 11% of tin. These filter materials are made from bronze powder having spherical particles closely graded in particle size, and this material is sintered in an atmosphere of cracked ammonia. Because the product averages a porosity of some 37% it has, of course, poor mechanical properties. The author cites tensile strengths of from 1 to 2 tons/sq. in. with about 3% elongation. These mechanical propertes, particularly in reference to resistance to shock, are, of course, very much superior to those of ceramic filter materials used for similar purposes.

The author cites some typical pressure-drop characteristics for these materials with different liquids and gases. These materials are being successfully used for the filtration of oil-engine fuel, for air filtration, and for the filtration of petrol and paraffin, and the author particularly refers to the treatment of water. They also have application for the distributing of liquids and gases, the mixing of liquids and gases, and as a flame trap. The paper is made more interesting by some good photographs showing the distribution of porosity in these porous products.

Another typical application of porous metal is described in the paper on aircraft liquid de-icing equipment by T.K.S. (Aircraft De-Icing), Ltd. The problem which had to be overcome here was the accurate distribution of very small and controlled amounts of de-icing fluid over the surface of an aircraft, coupled with the necessity to use a system which was mechanically robust, able to withstand the vibrations experienced, able to withstand high fluid pressures, and able to withstand the high mechanical stresses developed in aircraft skins.

The T.K.S. system consists of a tank containing the de-icing fluid, a porous filter to take dust out of the fluid, a pump capable of delivering up to 22 pints of fluid per hour, with a maximum pressure of 225 lb./sq. in., and sets of porous metal distributors made from a copper–nickel–tin alloy. These distributors are set in $\frac{1}{2}$-in. square section Monel-metal channels up to 7 ft. 6 in. in length. Capillary tubes set in the Monel-metal section containing the porous metal meter the flow of the de-icing fluid to an accuracy of $\pm 10\%$ per foot length of wing surface.

The authors of the paper point out that it was not convenient to try to control the rate of flow of the fluid by controlling the porosity of the porous metal, the main reason being that the amount of fluid flowing is very small indeed, which would involve making the porosity of the strip very fine, therefore rendering it very easy to clog up. The control of the flow of the liquid, therefore, is effected by these capillary tubes set at the back. The authors point out that almost all new British civil aircraft are now being fitted with this de-icing system, and it therefore

* In Table I, on page 100 of the first edition of the Symposium, the copper content of the type *A* bearing should read 87·20%, instead of 82·20%, and the tin content 10·57%, instead of 8·57%.

represents a new and unique British application of powder metallurgy.

Mr. **M. M. Hallett** (Sheepbridge Stokes Centrifugal Castings Co., Ltd., Chesterfield), opening the discussion on Section E, said: This Section of the Symposium is of particular interest, because it deals with applications which make use of the unique properties of porous metal components, and not merely with the use of powder metallurgy as one of the alternative means of reaching some desired end. Of the three papers, that by Mr. Sinclair in particular emphasizes the importance of the form of the porosity in the component. As powder metallurgists, we may sometimes be guilty of paying too much attention to the size and shape of the powdered metal particles and not enough to the size and shape of the pores which result from using certain particles. The size and shapes of those pores are obviously often very important factors in determining the properties of the products.

It has been proposed that a study group might be formed if the interest shown in this Symposium is sufficient, and I would suggest that one of the problems with which such a group might well concern itself would be the formulation of standard methods of expressing the characteristics of the pores in powder-metallurgy components. In the cast-iron world, a reasonably satisfactory method has been published by the American Foundrymen's Association and by the American Society for Testing Materials, which provides an easy method of expressing the size and type of graphite flakes in grey cast iron. A corresponding method, obviously differing in details, would be very useful in the powder-metallurgy world. That the parallel between grey cast iron and powder metallurgy is a real one, is well brought out in the paper by Chadwick and Broadfield in a rather unexpected way. The authors show in their Fig. 9 the results of quenching and tempering iron–graphite compacts, and it is very interesting to note that the hardened components have distinctly lower tensile-strength values after tempering at low temperatures than after tempering at some optimum temperature around 350° C. That is exactly what is experienced in the heat-treatment of grey cast iron, where 300–350° C. is also the optimum temperature. Probably the reason is the same in the two cases. The hard materials do not possess the ability to undergo slight plastic deformation to relieve the stress concentrations induced in one case at the ends of the graphite flakes and in the other at the roots of the pores in the porous metal components, whereas the slightly softer material, tempered at about 350° C., is capable of so doing.

Mr. **V. T. Morgan** (Messrs. Bound Brook Bearings (G.B.), Ltd., Birch Road, Witton, Birmingham): I would like to comment on the paper by Mr. Carter and Mr. Metcalfe. In their Figs. 1, 2, and 3, the authors reproduce photomicrographs of the three types of bearings which they examined. Comparing these with the corresponding density figures shown in Table I, type *A* has a density of 6·4 g./c.c. and type *C* a density of 6·2 g./c.c. These figures seem to be inconsistent with the reproductions. If the order were reversed, it would be easy to appreciate the differences in the porosity.

Care is needed when cutting a microsection from a porous bearing or a porous compact, to ensure that the plane being examined is at average density. I suggest, but I am open to correction on this point, that the authors have taken their section for the type *A* bearing, Fig. 1, from a portion of the bearing rather towards the centre, and that for the type *C* bearing more towards one end. If this is the case it is not strictly fair for the authors to conclude that the porosity of the type *C* bearing consists of many independent holes unconnected by capillaries. It is difficult, if not impossible, to judge the continuity of porosity on a two-dimensional plane; it is better to use some type of permeability test to investigate the nature of the porosity. The previous speaker suggested that there was a need for standardization in this form of testing, and that whereas a great deal of attention was paid to particle shape and size, not a lot was paid to pore size. If, in a permeability test, whereby either liquids or gases are caused to permeate through the walls of the bearing, the absolute viscosity of the medium that is flowing, the rate of flow of that fluid per unit pressure drop across the compact, and the dimensions of the bearing, are known, it is possible to calculate the absolute permeability. Also, by determining the porosity from the average density of the piece and the solid density of the base material and comparing this with the absorption of the piece, an estimate of the percentage of blind pores present can be obtained, and the average pore size determined by reference to the absolute permeability. From experience of this method of testing and comparison with the photomicrographs, one would expect that type *C*, far from having no capillaries, had in fact a higher permeability per unit density than the type illustrated in Fig. 1. This is an important point because it is on the basis of these photomicrographs that the authors have drawn conclusions in respect to the significance of their sintering experiments later on, the importance of the presence of the δ phase, and the anticipated running performance of the bearings.

In the discussion of results, on pages 101 and 102, the authors state: "The size and distribution of porosity are undoubtedly of great importance in the performance of porous bearings, and it is interesting to note that considerable difference was found in the bearings examined." Are we to understand that it is the performance in which they found the difference, or was it in the porosity? If it is the porosity, as shown in the photomicrographs, on which they judge the performance of these bearings, then the points already referred to regarding the position from which the specimen may have been cut and the need for a positive method for estimating whether or not the pores are intercommunicating, would have to be taken into account before arriving at a conclusion that bearing *C* would give inferior performance. A bearing with porosity similar to type *C* does not necessarily give a poor running performance, and it is not conclusive to criticize a bearing on the appearance of its microstructure. The type of porosity of the bearing frequently has to

be adjusted to suit the actual conditions of the application.

The paper by Mr. Sinclair shows permeability measurements in Figs. 9 to 13. With respect to Fig. 13, where air permeation is being used, it will be noticed that the pressure drop is not proportional to the rate of flow, whereas on the other graphs it is so. The rates of flow, when using liquids, are generally small, but with gases it is necessary to decide whether the pores behave as a series of parallel tortuous pipes or as a series of orifices.

When the flow is slow and viscous, the pressure drop is proportional to the rate of flow. As the pressure is increased, turbulent flow is induced and the bearing tends to behave as a compound orifice. In this case the pressure is proportional to the square of the flow, and a curve of the type shown in Fig. 13 is obtained. Owing to the compressibility of the gas, a Boyle's law correction is necessary, but this will not compensate for the disproportionality.

The sintering experiments conducted by Mr. Carter and Mr. Metcalfe, using electrolytic copper and 10% of tin, confirm actual production results, and the importance of the peritectic at 798° C. is well appreciated. The author's approach to the theory of sintering by X-ray diffraction methods is an interesting one, and as far as I know this is the first time that this method has been used.

It should be pointed out, however, that in practice there are other things which influence the sintering operation. Such factors as the density of the compact, the rate at which the pieces are heated, the properties, origin, and past history of the powders used, the time for which the specimens are held at the temperature, the atmosphere, and even the presence and nature of added lubricants are all possible variables, and can affect the sintering process to a marked degree.

The effect of the critical sintering temperature at about 800° C. has also been found to be dependent upon the amount of tin present. The higher the tin content the more rapidly the mechanism seems to proceed, and, whereas the authors have been working with 10% of tin, the actual bearings which they examined and to which they translated their research results, have in type *A* 8·57% of tin, in type *B* about 10·5%, and in type *C* 12·36%. The order of increasing tin content is also the exact order in which the authors found that these three bearings should be placed with respect to the quantity of δ-phase present. This may be a coincidence, but it is rather strange that on the one hand increased tin content causes an acceleration of the reaction, and on the other hand the quantity of δ-phase present was most in the case of the low tin content, and least in the case of the high tin content.

It has been found that the greater the sintering temperature and the higher the tin content, the greater the permeability. Yet that is the reverse order from that in which the authors have classified the three bearings.

The mechanism of sintering is neither wholly a question of whether the temperature is raised to above the peritectic at 800° C., nor as simple as the paper would lead one to believe.

Dr. **P. R. Marshall** (Ferodo, Ltd., Chapel-en-le-Frith, Stockport) : I should like to comment on the remarks made by Mr. Morgan. Referring to the graphs on pages 108 and 109 of the paper by Mr. Sinclair, only the line labelled " grade *D* " in Fig. 13 is curved. I suggest that the curve is due to the use of air rather than to the high pressures cited by Mr. Morgan. It will be noticed that the pressure axis of Fig. 13 is scaled up to 1·5 lb./sq. in., which is considerably lower than the pressures used to force oil through the pores. Incidentally, I once had occasion to calibrate for carbon monoxide and helium a flowmeter of the type which consists of a piece of capillary tubing with an oil manometer across it to show the fall in pressure. The calibration graphs were curved and that for helium lay above that for carbon monoxide, which is the opposite to what would be expected from published viscosity values. The discrepancy and the curvature of the graphs were finally shown to be due to my having further constricted the capillary tube near its centre, causing the gas to pass through the tube by some process between " effusion " and " streaming."

The determination of porosity by a liquid or air flow method is applicable to compacts which are to be used as filters or porous bearings. As a general means of determining porosity the diffusion method is not suitable for the following reasons : (1) The porosity may not be interlocking, and (2) the quantity of liquid which flows through the compact under a given pressure gradient is dependent on the distribution of the porosity, *i.e.*, whether it consists of a few large holes or a large number of small ones.

Mr. **P. D. Liddiard** (Glacier Metal Co., Ltd., London) : Section E of the Symposium represents a large proportion of the powder-metallurgy industry, but it is represented by three papers only. I think that bears out what Dr. Jones said in his opening paper ; this is a sphere in which very little progress has been made, so there is very little on which to report, apart from what has already been published over a number of years, largely from American sources.

One point which strikes me, not being a worker in the porous metal industry, is the difference between the types of powder used in these different porous objects. On the one hand are the porous bearings, which seem to utilize dendritic electrolytic powder for their copper content, and on the other hand are the porous filters, which use essentially spherical powders. The object is, I imagine, to have controlled porosity in each case. In the case of the filters it is controlled porosity to enable a given amount of liquid to flow through in a given time. I should have thought that to be one of the fundamentals of these bearings, in order to allow oil to be metered out gradually as required. It is necessary, therefore, for the amount of porosity and the type of porosity to be controlled.

Judging by the photomicrographs shown in the paper by Carter and Metcalfe—they have been subject to criticism, but they do indicate the type of porosity which is present—the type of porosity in this case is extremely irregular both in size and in shape, whereas the type of porosity in the filters,

as indicated by the photomicrographs in Sinclair's paper, is certainly far more controlled both in size and in shape.

This question of the size and type of porosity has been raised by several speakers, also the methods of measuring or observing it. I would suggest that we should always bear in mind the ultimate use of the products. We do not necessarily want to examine the types of porosity just to see what they are like ; we want to know what effect a given type of porosity is going to have on the ultimate use—the amount of oil which it will retain or pass, and so on. Our ultimate test might be designed round some factor of that kind—the amount of oil which it will retain and which it will pass in a given length of time. One method of determining the difference between interconnected and non-connected porosity would be by differences in density determination with and without the presence of oil.

I feel that the density figures given in the paper by Carter and Metcalfe would be more useful if they had been given after the extraction of oil. I imagine that the figures shown in Table I are with the presence of oil. If they had been given without oil we could obtain some indication of the degree of porosity, having a knowledge of the densities of the fully compacted material. We are given the compositions, and it would be possible to find out the densities of the fully compacted materials, and from that to get an idea of the ultimate density and porosity of the bearings in question.

I agree with Mr. Morgan that all this work has been done around a very critical composition as regards tin and copper. One has only to look at the equilibrium diagram to realize that variations in the tin content and in the conditions under which the two metals were brought together and maintained at elevated temperatures, and in the conditions under which they were cooled, will make a very big difference to the ultimate structure as observed under the microscope or by X-ray examination.

Dr. **C. H. Desch,** F.R.S. (President of the Iron and Steel Institute) : In connection with the determination of the total porosity and the interconnected porosity in a mass of this kind, I would call attention to the methods used in the cement industry, where you have the same problem. It can be done easily by using a liquid which penetrates readily. With cement xylene is always used, which penetrates readily where there are connected pores ; and then, by comparing the amount of liquid taken up with the known density of the material, obtained either from knowledge of the composition or from the determination of the density of finely crushed material, one can obtain both the total porosity and the interconnected porosity.

Dr. **L. Northcott,** Mr. **C. J. Leadbeater,** and Mr. **F. Hargreaves** (Armament Research Department, Woolwich) wrote : In his remarks upon the non-agreement of the figures given for the densities of three types of bearing, *A–C*, and the photomicrographs of their structures (Figs. 1–3), Mr. Morgan appears to have overlooked one salient point. The porosity of a sintered material is dependent not only upon its actual density but also upon the theoretical density for the material and therefore on the composition. The values of the latter quantity for types *A* and *C* have been calculated as 8·462 and 8·174 g./c.c. respectively. As the actual densities of the bearings lie in the ranges 6·35–6·44 g./c.c. (6·4 g./c.c. is the figure quoted by the authors) and 6·15–6·24 g./c.c. (6·2 g./c.c. is quoted) respectively, the corresponding porosities lie in the ranges 24·96–23·90% and 24·77–23·70%. The discrepancies between the photomicrographs and the density values are thus not nearly so large as Mr. Morgan is led to assume.

The values of porosity range quoted above (1% in each case) also serve to show the care needed when density values are approximated.

AUTHORS' REPLIES

Mr. **A. Carter** (Powderloys, Ltd., Coventry), in reply, said : This discussion has at least produced some very interesting points which have not been raised before on the subject of porous bronze bearings.

Dr. W. D. Jones seemed to me to make a point—I am not sure that I heard him correctly—with reference to the different rates of attaining equilibrium shown by mixtures containing fine and coarse tin. I think that it is natural to expect that compacts containing coarse tin will reach equilibrium rather more slowly than those with fine tin, mainly owing to the greater degree of intimate contact obtainable when using a finer tin powder and to the shorter distance over which diffusion has to take place. It must be remembered that these compacts were sintered for only a comparatively short time at temperature. For instance, the sintering speed mentioned in the original experiments was 1·67 cm./min. This speed is such that the compacts have approximately 1 hr. in the total length of the hot zone of the furnace, but only 10–15 min. at the maximum temperature.

Dr. **W. D. Jones** : The paper points out that the coarse compacts reach equilibrium more quickly.

Mr. **Carter** : I am sorry ; that may be misleading. The point is that equilibrium is reached at a faster rate with the compacts containing coarse tin, but the temperature at which the solution becomes rapid is *higher* for the coarse powder, and is related to the temperature at which much liquid appears. Comparing the structures at the same temperatures, *e.g.*, 700° and 750° C., the compacts containing fine tin are more homogeneous than those containing coarse tin. This is what we meant by saying (on page 104) that " the compacts containing coarse tin powder suddenly reach equilibrium above 800° C., whilst the fine tin compacts gradually approach homogeneity."

Mr. Morgan raised some very interesting points, the first being in connection with the microstructures. We admit that visual examination of the microstructures will suggest that sections of the bearings which are shown in Figs. 1 and 2 are less dense than that shown in Fig. 3, but we must emphasize that it is the *type* of porosity which we are examining, and not the actual amount. I am afraid that I do not remember from just which part of the bearing these photographs were taken. The estimation of the amount of the porosity of bearings simply from a

two-dimensional photograph is very difficult. We had intended to pursue some permeability experiments on these bearings, but these have not yet been done.

I would emphasize that it is not only the permeability of the bearing which is of importance, but also the ability to retain oil in the bearing. We consider that the oil is retained by the bearing in both the blind pores and those communicating with the surface. These blind pores act as reservoirs and supply oil to the bearing surface through the capillaries. We must also emphasize that the work reported here was intended only as a primary study of the subject of the sintering of these bearings, and we agree with Mr. Morgan that so many other factors are involved in the sintering, such as the action of volatile agents in the compacts, the grain-size of the copper and tin powders, all of which have a considerable effect on the structure of the bearings, that all these points have to be borne in mind. We note that in general, however, Mr. Morgan finds that a change of physical properties usually occurs at or about 800° C.

In our discussion of results on pages 101 and 102 we did not intend to imply that we have conducted any running tests on these bearings. We did no bearing tests and we could not find any literature relating to any bearing tests having been done on porous bronze bearings, apart from a few statements on the performance and wear of individual bearings under certain conditions. If any work has been done it should be published, and if it has not been done it should be done. We feel that this lack of information applies to the whole field of porous bronze bearings, and this may be due to the fact that many of the processes are covered by American patents. There must be a considerable amount of information of general interest to those engaged in powder metallurgy, which could be published.

Mr. **C. E. Sinclair** (Sintered Products, Ltd., Sutton-in-Ashfield) replied: The most important property required in porous materials used for filtration, aeration, and the distribution of fluids, has been indicated by Mr. Hallett, namely, the shape and type of porosity. With only a limited knowledge of porous bronze bearings it appears to the author that in porous metal of this type it is not essential for every pore to be of uniform size, but when a material is to be used for filtering a liquid, or to produce a definite bubble size as an aerator, every pore should be continuous and as nearly the same size and shape as possible. Experiments have proved that this can only be achieved by using powders with spherically shaped particles.

Early work done with powders containing irregularly shaped particles in the production of highly porous compacts showed that it is impossible to produce uniform grades of material by the normal methods of screening powders. The shape of the powder particle, rather than its size, has more effect on the cavity in the compact. Experiments showed no uniformity in flow through different samples of the same nominal grade, owing to the irregular shape of the pores.

Interesting light has been thrown on the curve shown in Fig. 13, but the following point should be borne in mind when studying all these graphs. They are drawn from experiments made on production material and not on specially prepared laboratory samples, and in each case an average of a considerable number of results has been taken. Variations in results are obtained with different specimens of one grade taken from production owing to the fact that it is not economical to screen powders to fractions close enough to produce an ideal material with uniform pore size. The variations are insignificant in practice, but in some cases make it difficult to determine whether a graph should be straight or curved. It is possible, therefore, that some of the other graphs should be curved as is shown in Fig. 13.

Captain **R. R. Rawson** (London), replying to the discussion on the paper by T.K.S. (Aircraft De-Icing), Ltd., said: The problem of de-icing is rather far separated from that of powder metals, and I am not an expert on that side of the subject, but there has been some discussion on the question of powders and filtration. When looking for a satisfactory material for the distributors, we were faced with this question of getting uniform flow, and at the time we had to depart from the idea of getting control by the porous metal itself on account of the difficulty—mentioned by Dr. Jones—of having very fine pores which became clogged up with dirt, and also because of the difficulty of ensuring that the required flows were kept constant under manufacturing conditions. It would have helped us, probably, if we could have had spherical powders, but at that time they were not available, and I understand that even now they are not available in the materials which we find it necessary to use. We had therefore to depart from using porous metal alone, and had to use capillary tubes, so that variations in the porous metal had little effect. The pressure drop through the porous metal is less than 0·05 of the total drop, so that slight variations in the porous metal have very little effect on the overall efficiency of the distributors themselves.

The Chairman proposed a vote of thanks to the authors and to all those who had taken part in the discussion.

Mr. **Carter** and Mr. **Metcalfe** wrote in reply: We were very pleased that so many contributions were made to the discussion, but we agree with Mr. Liddiard when he comments on the small number of papers which have been presented on a subject which represents a large proportion of the powder-metallurgy industry. On the other hand, we do not think the reason can be solely that little progress has been made, in view of the considerable amount of unreported experimental data which has been mentioned in the discussion. We do feel that there is a considerable amount of such data which should be published in order to help solve some of the more fundamental problems connected with the process of sintering. For this reason we rather deplore the apparent lack of interest shown by some of the manufacturers in this field.

The importance of the presence of a liquid phase was noted by Dr. Jones, and this would suggest that a marked decrease in the rate of homogenization should be observed when the liquid disappears as a result of diffusion. This conclusion may only be valid below 798° C., because at this temperature the liquid phase suddenly increases in amount, becoming, in some

cases, a stable phase. The presence of liquid introduces the possibility of mass movement which, together with the much higher rate of diffusion, would give rise to a rapid attainment of equilibrium, followed by a period in which continued heating has little effect on the final structure. The similarity of the pieces sintered for 10 min. and 70 min. at 750° C. (Table VII—a stoking speed of 1·67 cm./min. gives the test-pieces about 10 min. at maximum temperature) suggests that the liquid phase has disappeared in less than 10 min. when this rate of heating is used. The amount of liquid phase may be augmented by increasing the tin content and this probably accounts for Mr. Morgan's statement that an increase in the tin content increases the rate of the process. It would appear therefore that the use of a higher tin content would have given similar results except that, for higher tin contents, the same state of equilibrium would be reached at a somewhat lower temperature and *vice versa*. Above 798° C. an increase in the tin content would have the same effect as an increase of temperature, *i.e.*, to increase the amount of liquid present. The effect which we observed at 900° C. (*see* Table IV and Fig. 4 of our paper) might be more pronounced, however, since the first solid to be deposited from the liquid would be further removed from the equilibrium composition of the alpha phase if the tin content were higher.

Dr. Jones has pointed out that the behaviour represented by Figs. 4 and 5 is rather strange, and we would like to explain these figures in greater detail. When specimens containing fine tin powder are heated up, they gradually approach equilibrium, becoming homogeneous above 800° C., whereas those containing coarse tin powder are little affected below 750° C., after which they rapidly begin to approach the homogeneous alpha phase. Thus equilibrium is reached faster, in its final stages, with coarse tin, but the temperature at which the solution becomes rapid is higher, and is related to the temperature at which much liquid appears. Comparing the structures at the same temperatures, *e.g.*, 700° C. or 750° C., the components containing fine tin powder are more homogeneous than those made from coarse powder, as would be expected.

Above 850° C. the fine-tin-powder compacts show an increase in the range of composition of the alpha phase, whereas the coarse-tin-powder compacts do not. The explanation of the broadening of the range of composition is that at 900° C. a 10% tin alloy in equilibrium would contain nearly 70% of solid alpha phase (6% tin) and 30% of liquid containing 19% of tin. The solidification process would then correspond to that in the casting of a bronze, resulting in cored structures. The absence of such a broadening in the case of the coarse-tin compacts may be due to the higher temperature at which a uniform alpha phase first appears, but the exact mechanism is not clear.

Mr. Morgan mentioned various factors which affect the sintering operation, and of these, the compact density, sintering time, and temperature were investigated. In our experiments the rate of heating varied with changes in stoking speeds, and since these varied by a factor of 3, we consider our results indicate that the rate of heating is not critical over the range investigated.

The paper is mainly concerned with the structure of the type of alloy used in bearings of this type, and although permeability and bearing tests had been considered, this stage of the work had not been reached when the paper was published. Thus in our discussion on pages 101–102 of the Symposium, the differences refer to the size and distribution of the porosity, and as Mr. Liddiard points out, the photomicrographs show the type of porosity which is present, and were included solely to show this. Apart from a few statements on the performance and wear of individual bearings under certain conditions, there appears to be no literature on the subject. Although we have not done any bearing tests we found it convenient to adopt the following theory for adjudicating the probable performance. It was considered that oil would be retained both in the blind pores and the interconnected holes, which act as reservoirs, whilst it is necessary that there should be a series of fine capillaries to supply a regulated amount of oil to the surface. This was held only as a tentative view, and although we now feel it to be inadequate, it does seem that the measurement of permeability is not the only criterion by which the bearings should be judged. It would be very interesting to know if any bearing tests have been done which would show whether this view is correct or not.

The density figures were given after the oil had been removed by several treatments in hot carbon tetrachloride. The difference between the measured density and the appearance of the microsection is no doubt due to the use of a section of which density is not the average of the whole, as suggested by Mr. Morgan. In the case of type *C* bearing, we have carried out further tests and the results are shown in the data given below, and in Figs. D–H. The following data show the hardness traverse for this bearing, which was about 4 in. long. Considerable variation exists, and Figs. D, E, and F show the porosity at distances of $\frac{1}{4}$, 1, and 3 in. from the harder end (diametral plane), while Figs. G and H show the porosity at distances of $\frac{1}{4}$ and 3 in. from the same end (axial plane).

Hardness Traverse

Distance from End, cm.	Hardness, V.P.N.	Distance from End, cm.	Hardness, V.P.N.
0·43	63·1	5·39	52·0
0·94	64·3	6·79	49·8
1·61	66·8	7·76	48·9
2·49	65·3	8·70	57·9
3·42	60·6	9·10	52·2
4·37	53·8		

The actual difference in the sections is small, but it does show that the type of porosity is similar throughout the bearing. The axial sections show that the porosity is very similar in two planes at right-angles, from which it may be taken that the type illustrated by one section is representative.

The difference in density along the length of the bearings is of course due to the difference in pressure along the length of the compact. This may lead to differences in the performance of the bearing.

Finally, we would like to express our thanks to all those who contributed to the discussion in the course of which many new facts have been brought to light.

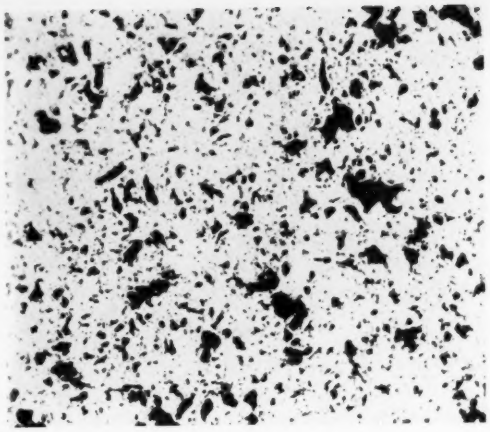


FIG. D—Diametral plane, ¼ in. from end

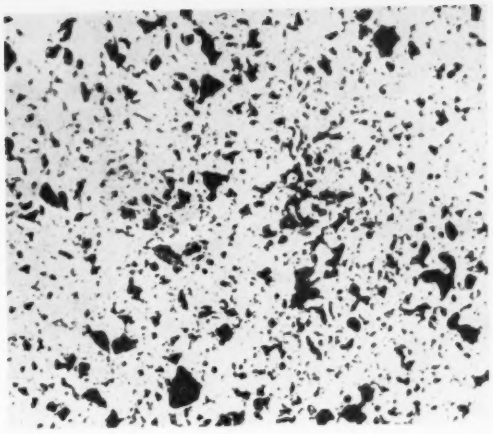


FIG. E—Diametral plane, 1 in. from end

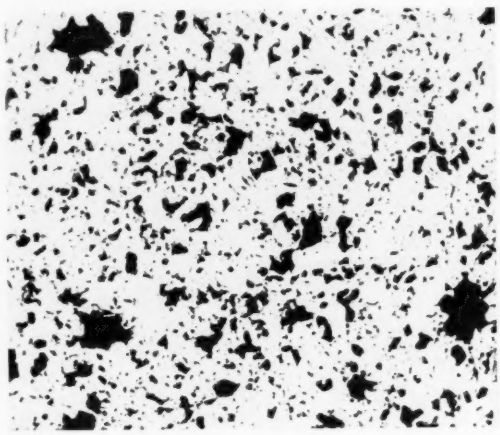


FIG. F—Diametral plane, 3 in. from end

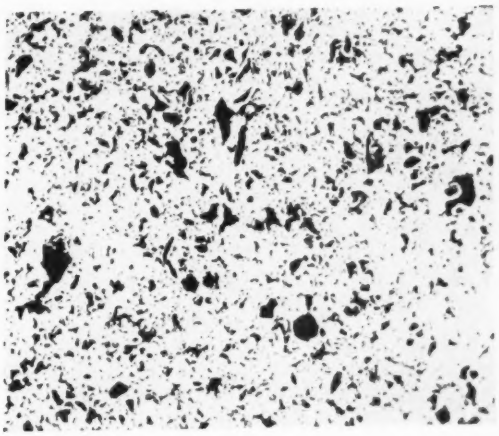


FIG. G—Axial plane, ¼ in. from end

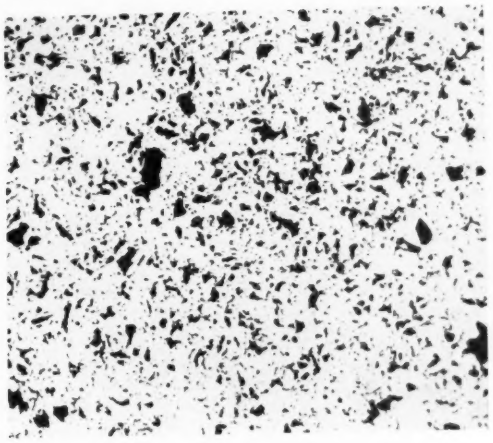


FIG. H—Axial plane, 3 in. from end

FIGS. D to H—Porosity of type *C* bearing × 50
(*See* Carter and Metcalfe's reply)

Discussion on

Section F—THE MANUFACTURE AND PROPERTIES OF SINTERED COMPONENTS

Part 1—Ferrous Components (Based on Papers 18 to 22, 27, and 28 of the Symposium)*

The **Chairman** (Lord Dudley Gordon) said that Mr. J. A. Judd was to have acted as rapporteur for this Section of the Symposium, but, as he was on the point of leaving for Australia, Dr. Northcott had kindly consented to act in his place.

Dr. **L. Northcott** (Armament Research Department) (rapporteur), presenting this aspect of Section F of the Symposium, said : Of the seven papers which I have been asked to introduce, the first, by Mr. Burr and Mr. Clarke, is the only one dealing with straight iron powder compacts in simple form. These authors found that when a compact, and particularly a long compact, is pressed, the compacting pressure would appear to be different at different parts of the specimen. By preparing a series of specimens of different lengths, compacted under different compacting pressures, they found that the compacts so prepared had different final densities. By plotting the average density against the length of the compact so prepared, they can extrapolate back to a limiting length of compact of 0, giving a limiting density—that is, the maximum density in the compact adjacent to the top and bottom plungers. From that they have estimated the density also of the centre and less dense part of the compact, and have determined the loss in pressure due to the die-wall friction, and so on, exerted.

The next two papers, one by Mr. Judd and one by Mr. Chadwick and Mr. Broadfield, deal with iron–carbon alloys and their preparation by powder metallurgy. Mr. Judd starts off by describing briefly the three methods for making iron–carbon compacts—that is, steel-type compacts. In the first instance, it is possible to start off with steel-type powders. The disadvantage of this method is that the compacting properties of steel-type powders are poor. Alternatively, by the second method one can start off with a soft iron powder and mix graphite with it, and at temperatures of 1000° C. or thereabouts the carbon goes into solution, and on cooling a steely type of material is obtained. The disadvantage of that process is that the material is very susceptible to decarburization, and also to lack of uniformity in carbon content throughout the compact.

The third method is to use a soft pure iron and introduce the carbon mainly in the form of an iron carbide, which either melts or approaches the melting point at the sintering temperature. If necessary, further carbon can then be added in the form of graphite. Using that third technique, the author goes on to discuss the preparation of sintered iron piston rings, and I think that these comments will be read with considerable interest, since piston rings are difficult to make by any method. Although obviously there is no such thing as a perfect piston ring, by this method the author seems to have attained in practice at any rate a very close approach to what is wanted in piston rings. One advantage of using his third method is that by using an alloyed carbide it is possible to introduce alloying constituents into the compact fairly readily.

The third paper, by Mr. Chadwick and Mr. Broadfield, deals with the pressing, sintering, heat-treatment, and properties of iron–graphite powder mixtures. These authors examine the properties of compacts prepared by using the pure-iron-plus-graphite technique. They start off with three varieties of iron, an electrolytic-iron powder, a Swedish powder, and a German powder made by the Degussa process. Their compacts were in the form of small rings approximately 1 in. in external dia., 0·1 in. in wall thickness, and 0·2 in. in height. These were sintered in an atmosphere of cracked ammonia, and subsequently examined.

In Table V, page 127, the authors give an interesting tabulation of the conditions affecting decarburization, and find that nil mean carbon loss in their experiments was obtained only by the addition of propane to the furnace gases, but a close approximation was obtained by partial protection from the furnace atmosphere. This is the practice used, I understand, in the greater part of their paper. They look fairly closely into the effect of the added graphite and the effect of sintering temperature and sintering time, and the results are plotted in the form of curves in Figs. 4 to 7.

The hardening and tempering technique is then discussed, the results being indicated in Figs. 8–12. The authors found certain anomalous results in the mechanical properties after hardening and tempering, and at first concluded that the anomalies were due to the variation in porosity. In order to obtain some information on porosity, they carried out tests on the permeability of the sintered compacts. It would appear that the reason for occasional failure to harden with insufficiently drastic quenching was the result of the lower heat conductivity of a porous material relative to that of a solid material. The authors go on to say that for suitable quenching the material must be sufficiently porous to enable the quenching medium to permeate through the compact, to exert a quenching effect, or it is necessary to adopt water-quenching ; and the disadvantage of water-quenching with a porous material is that it tends to break up the material.

There follow two papers on iron–copper compacts, one by investigators of the Armament Research Department, with which I shall deal first. These authors studied the properties of sintered compacts containing additions of copper to iron. They examined a variety of irons and a variety of coppers, and the influence of the copper content is shown briefly in Fig. 1 on page 143, which shows that maximum

* In Figs. 2 (*a*) and 2 (*b*), on page 145 of the first edition of the Symposium, the captions to the broken curves should read : " No. 9, 20% Cu (12 μ)."

XUM

tensile strength is obtainable with about 20–25% of copper. They examined the sintering temperature only in respect of whether the sintering temperature is above or below the melting point of the copper, and they show that it is desirable for optimum effects to maintain the temperature at about the melting point of copper ; *i.e.*, to ensure that the copper is in the form of a liquid phase.

The difference in the properties of these compacts, which were in the form of tensile blanks 6·5 cm. × 1 cm. square, is shown in Fig. 3. The microstructures with sintering temperatures of 1100° C. and 1050° C. are shown in Figs. 5 to 10. All the results which these authors obtain have been plotted in Fig. 16 to show the influence of copper on alloys of equivalent density, and they show that, for a given tensile strength, it is possible to prepare alloys of different porosities and copper contents, but that in each curve the peak for the maximum tensile strength occurs at about 10% of copper.

The next paper is by Chadwick, Broadfield, and Pugh. As in the case of their iron–graphite paper, these authors use compacts in ring form. Certain of their results are very similar to those in the previous paper. The maximum tensile properties are shown at a copper content in the region of 20–25%. The authors show some interesting curves in Fig. 3, demonstrating the effect of variation of particle size, and give other curves showing the effect of different processing conditions.

The two other papers which need to be considered here are that by Mr. Leadbeater on page 191 and that by Lt.-Col. Ivory on page 203. Both of these authors have spent an appreciable time in Germany investigating the position of powder metallurgy. Mr. Leadbeater gives a very interesting table (Table I on page 193) showing the production of metal powders. The only individual figures to which I would draw attention are those in the penultimate column, showing the total iron production. It will be seen that in 1944 over 32,000 tons of iron powder were prepared in Germany, and the projected production for 1945 was about 56,000 tons. In the earlier stages of the war, the Hametag process was the one mainly used, but later on the atomization methods, the Disc process, and the *RZ* process, were coming into some prominence.

The only other point in this paper to which I would draw attention is the comment which the author makes about the use of multiple-die punches. The question of die design has been gone into very thoroughly in Germany, and it is quite possible that we may benefit to some extent from the experience of the Germans during the war. The paper concludes with a useful appendix in the form of an index of references to powder metallurgy in various B.I.O.S. and C.I.O.S. reports.

Lt.-Col. Ivory's paper starts off with a review of the methods of making the powders and the types of powder produced, and he goes into some details which are of value and interest to some of us, but which are only of passing interest to most of those present, about the question of components for armaments.

Mr. **D. A. Oliver** (Director of Research, the B.S.A. Group of Companies), opening the discussion, said : I think that the papers in this section are very stimulating, as showing that the scientific foundations of the powder metallurgy of small steel and non-ferrous components are now being firmly and surely laid in Great Britain. In the past there has been a tendency to rely on secret empirical work, and I should like to emphasize that we are now beginning to publish the basic work and to help one another to arrive at a scientific understanding of these processes.

From the engineering angle, there is one thing that I should like to point out. In the production of an engineering component, it is going to be a difficult task for the powder-metallurgy method to supersede the older methods unless applications can be selected where very large numbers of a component are needed on strictly standardized lines, because the cost of tooling-up most components is relatively high. The plant overheads are also high if the plant is not fully occupied, which means that an effort must be made to try to even-out the load on a given plant. If that can be done, and if we can select engineering components which are wanted in vast quantities—perhaps millions—there are very good hopes for the new method. The papers which we have had presented are going to help us in securing higher strengths with other properties that are worth while.

It appears that the steels made from powders in Germany had advanced to a fairly high degree of technique, but on an experimental footing only. In America I think there have been greater commercial advances, but the properties in the components are inferior to many of those recorded in this Symposium. The position was aptly summarized recently by the statement that the American powder steels are generally comparable in strength with ordinary cast iron.

We are now creeping up in strength properties, and I am sure that the powder metallurgist will shortly be able to offer properties which are extremely competitive with the properties of simpler materials directly machined or precision cast. We have therefore three methods : The direct moulding from powders, the precision-casting method, and the orthodox direct-machining method. These three methods, I submit, should now be considered side by side, always provided that components can be selected which lend themselves to mass-production and which, in most instances, can be severely standardized.

Mr. **R. J. Brown** (Nuffield Central Research Laboratories) : As a prospective user of these various materials and components produced by the method of powder metallurgy, I should like to say a few words. The application of powder metallurgy to the production of carbide tools, sintered bearings, ferrous bearings, and filters, has been well established for a great many years, and there is no doubt that such components adequately meet a vast need and show a considerable advance over the older materials ; but when consideration is given to the use of sintered ferrous components the picture is somewhat different. I have personally had the somewhat unfortunate experience

of endeavouring to substitute a sintered iron–graphite component for a cast-iron component with astonishing results ; the sintered component actually disintegrated in use with little or no evidence that it had ever existed. The disintegration of these components appeared to be related to a considerable variation in density—density as between the component and density within the individual components. At the time, this was stated to be due to the poor flowing properties of the powder available, but examination of numerous components of various types produced by powder metallurgy indicates that this density variation is a problem which has not yet been overcome. It has already been stated in the discussion that density will vary from end to end of the sintered compact, and presumably from the surface of the cross-section to the midway position in the section.

Referring to a remark made by Mr. Oliver, that the quality of American sintered ferrous components is equivalent to that of cast iron, I assume that his implication is that by comparison sintered ferrous components produced in this country resemble steel ; if I am correct in this assumption, I can only voice a wish that the best of the sintered ferrous components produced in this country were equivalent to cast iron, as they would then find many applications in the engineering industry. Unfortunately even the American components do not compare favourably with cast iron.

Many references are made in the American technical press to the greatly improved mechanical properties obtainable in sintered ferrous components by either hot-coining or hot-pressing, or by a combination of both processes ; the properties quoted certainly approach those obtainable in steel produced by conventional methods if tensile strength and elongation are accepted as the criteria, although impact values are lower. The need for hot-pressing or hot-coining is vital if sintered ferrous components are to be considered as a prospective competitor of the conventional ferrous metal.

Referring to the paper by Northcott and Leadbeater, in their discussion on page 149, the authors say " the mechanical properties of sintered compacts are dependent upon (1) the cohesion between the individual metal particles, (2) the strength properties of these particles, and (3) additionally in the case of 2-phase compacts, upon the cohesion between the primary particles and the second phase." Too much emphasis cannot be laid upon this property of cohesion, because it is obvious that it is lack of cohesion between the individual particles which causes the mechanical properties of sintered iron compacts to be so poor. It is essential that the metallurgist associated with this new metallurgical development pays attention to the development of satisfactory strength and adequate ductility in his products, and he cannot afford to overlook impact strength whether it be measured by the much maligned Izod test, probably modified as by the British Cast Iron Research Association, or by the engineers' practical test with a hammer.

Dr. **W. D. Jones** (Consulting Metallurgist, London) : I cannot undertake to answer the points which Mr. Brown has raised directly, but it so happens that I was intending to pass a few comments on some of these papers relating to matters which he has already brought up. Mr. Brown is correct in saying that bad density distribution is one of the current defects of the products of powder metallurgy. I should like to draw attention to a paper from the school of Professor Wulff, which was published as a technical publication of the American Institute of Mining and Metallurgical Engineers, on the subject of plastic deformation in metal-powder compacts.

A study of the curves in that paper shows quite a remarkable fact, and that is that if the die is lubricated a much higher density and a very much better distribution of density across a component is obtained than if the powder is lubricated by adding a lubricant to it. The author of that paper actually shows that there is hardly any point in adding a lubricant to the powder, and in fact it is perhaps a bad thing, because the lubricant does interfere to some extent on some occasions with the sintering properties of the powder ; while by lubricating the die a better result is obtained in any case.

I bring that point up not so much to suggest that Mr. Burr and Mr. Clarke might have made reference to Professor Wulff's work, but rather to draw particular attention to this paper, because it seems to me to be a very important one, which has so far tended to be overlooked. I feel that if the die manufacturers or press manufacturers in the powder-metallurgy business were to give some attention to the possibility of lubricating the die automatically we might get better products and not have to worry about adding lubricant to the powder. It does seem to me to be a very difficult proposition to arrange a press in which the die set is automatically brushed out with lubricant once every few cycles ; it probably means the development of special press attachments.

Going on to the paper by Mr. Judd, Mr. Brown suggests that possibly one of the steps towards getting a more satisfactory component is to use hot-pressing. The remark was made in the discussion this morning that hot-pressing had received little or no attention in this Symposium, and that is quite true. Mr. Judd refers to hot-pressing on the second page of his paper, and in the second column of that page is kind enough to refer to some results which I myself obtained by this technique.

I have always been an enthusiast for hot-pressing, and so far I have been a little disappointed that it is not commercially used to a greater extent. I should, however, like to join issue with Mr. Judd—I am sorry that he is not here to make the junction—on the remarks which he makes about hot-pressing. He points out that high tensile strengths and high densities have been obtained by the hot-pressing of powders in air. He suggests that the results may be partially due to decarburization because no precautions were taken to provide a suitable atmosphere, and then he says " This need for the provision of a non-oxidizing atmosphere . . . is the obvious commercial limitation of the hot-pressing technique." That, I think, is exactly where he is wrong. The hot-

pressing work to which he refers was done deliberately in air, because it was found that much better results were obtained in air than in a protective atmosphere. The results were not due to decarburization, because very excellent results have also been obtained with brasses and bronzes, and, working with 90/10 bronzes, it has been possible to obtain hot-pressings with an elongation of the order of 75%, which is quite a remarkable figure.

I think that the hot-pressing technique needs to be studied more closely to get good results, but, of course, the real difficulty is not the question of protective atmosphere but a die difficulty. That has already been discussed in connection with the carbide papers. If pressing temperatures of the order of 800°, 900°, or 1000° C., are used, so far the only die materials which appear to be available are materials of the order of graphite, which have a very poor tensile strength and which seriously limit the pressures which can be used. Moreover, of course, such materials wear very quickly, and it is not possible to keep within close tolerances for very long.

I wonder whether, in studying hot-pressing, we have not perhaps overlooked the possibilities of techniques which are used in other branches of metallurgy, for example, drop-forging. I myself am of the opinion that the best method from an industrial point of view in producing large quantities of hot-pressings cheaply would be perhaps to press the powder cold and heat the compact as quickly as possible to the sintering temperature, which perhaps can be best done by induced high-frequency current, then forge the hot compact in a cold die. That, I think, gets round the difficulty of the hot die. Work has been done along those lines in a rather crude way by several people, and I understand that in America the Chrysler Company did fashion some steel components in much that way. I think that it needs elaboration, and I am sure that it does not need a protective atmosphere.

I should like to refer to a point in Mr. Leadbeater's paper, where he refers to the D.P.G. process. In other papers the same process is referred to as the Disc process, and it is also referred to as the comminution process in one of the papers. I should like to amplify a little what is said about this process. It was developed about 1937 or 1938 by the Deutsche Gold U. Silbescheideanstalt in Frankfurt, and arose out of one of their methods of comminuting or granulating chemical materials. It consisted then of an aperture through which molten metal was poured, a fairly large diameter cone of water, and a rotating wheel with radiating spokes or knives on it which rotated at a high-speed—4000–6000 r.p.m. The cone of water was brought more or less to a focus on this rotating wheel, and the technique involved was a breaking-up of the molten metal by the moving knives and the immediate freezing of the particles by the surrounding water.

I should like to add what I may describe as one or two home truths about the machine, because it may be of interest to mention that two of these machines were working in this country during the war. It happened in this way. Dr. J. E. Hurst, (of Messrs. Bradley and Foster, Ltd.) and I went to Frankfurt, and came back with drawings of this machine a few days before war commenced. One of these machines was installed quite quickly in London, and one in Darlaston, the latter working on iron and the London one on non-ferrous metals. As far as I know, the Germans made very few changes in the machine after we obtained the drawings, and I should like to point out that both these machines were extremely unsuccessful. There were many difficulties with them. One acute difficulty was the fact that molten metal was thrown on to the rotors and built up on them, so that the machine had to be stopped at frequent intervals, perhaps after every few hundred pounds of metal had been poured, to get the frozen metal off the rotors—if it did not come off itself in one large lump and go through the casing. Further, the powder produced by these machines had a very high oxygen content, generally of the order of 1–3%, and was very coarse. These machines were eventually replaced by atomizing methods, which were considered to be an improvement.

The question arises of why the Germans went on using these machines when they were unsatisfactory. Mr. Leadbeater and others point out that they were unsatisfactory in many ways. There were large numbers supposed to be working, of which a considerable percentage at any time had broken down. It is also clear that the powder produced by these machines was invariably reduced before use, which was the way in which they got round the high oxide content. I imagine that the German technicians, being pressed for a large production of powder, quickly seized upon a method which gave reasonable results and just duplicated those machines, made as much as they could, and then took their time towards the end of the war to consider improved methods. Perhaps Mr. Leadbeater will be able to say whether similar remarks could be made about the Hametag process. I have noticed the growing interest everywhere in the D.P.G. process, and it was with the idea of telling these home truths about it that I decided to make these remarks.

Dr. **C. H. Desch,** F.R.S. (President of the Iron and Steel Institute) : I am going to confine myself to the subject of the iron–carbon alloys, because I can see the great advantages which can be obtained with other alloys by using this method of preparation ; but in the case of the iron–carbon alloys, after looking through these papers I am still not convinced of what is to be gained when the aim is not to produce a porous material but to have a solid material.

With regard to the cost, compared with machining there is naturally an immense saving, but it is also necessary to compete with the method of precision casting, which is gradually being improved and which, of course, gives a product very greatly superior in mechanical properties to anything which can be prepared by powder metallurgy. We have here a fairly good tensile strength, but very poor elongation as a rule. When we hear that these things can be improved by forging the powders under a press, we cannot but wonder why we should not start with an ordinary steel to begin with, if it is necessary to

go through such a complicated process as that.

In March last I was in the U.S.A. and visited the big Department of Powder Metallurgy at the Stephens Institute at Hoboken. I saw the results obtained and brought away some of the specimens, which are of about the same quality as is described in these papers.

I notice that there is very little reference to the use of iron powder prepared by the carbonyl process. The one paper which mentions it says that it is relatively hard and costly, but I found that at Hoboken they were using mainly carbonyl iron. It has the advantage of having spherical particles, and I am surprised at the statement that it is hard, because my experience is that it is soft. These spherical particles give a very compact mass when packed together, and in preparing alloys from an iron powder it seems to me that this would be the ideal powder for the preparation. I cannot say anything on the question of cost. No doubt the carbonyl process is expensive, but it was extensively used in Germany, and even sheet and plate material was prepared there from sintered carbonyl iron, giving a good product. When it is necessary to have a porous product this is an admirable method, but I am not convinced from these papers of the value of the method for making steel.

Mr. L. J. Brice (Ministry of Supply) : I was interested in the remarks of Dr. Jones regarding the D.P.G. process. He is to be congratulated on having been able to extract documents from Germany and get them back to this country so rapidly.

Recently I had an opportunity to visit an eminent Austrian metallurgist who is now associated with an Austrian firm in the development of this same process, and I was informed that they are already producing two tons of steel powder per day. It was alleged that details of this process were unknown in this country.

Research is proceeding in this firm on the production of rolled and drawn products prepared from the powder, and already the production of wire of a quality suitable for use in mine haulage ropes has been shown to be practicable, though so far only on an experimental scale. Welding rods are being produced from the powder, a speciality being rods in which the flux is integrally mixed with the powder instead of being used as a coating or a core. The firm is also producing sintered products. I was informed that projected new developments are in the direction of producing alloy-steel compacts, produced from compression of the normal steel powder mixed with the necessary alloy components.

If anyone is interested in these developments, the Ministry could probably arrange introductions for ordinary commercial negotiations. Owing, however, to the special conditions which apply now to Austria, it is not possible to divulge details of the firm's activities in the way that information from Germany can be given through B.I.O.S. channels.

I would like to refer to two or three B.I.O.S. reports which have been made available since the preparation of the list given in Mr. Leadbeater's paper on page 200. The report of the team which investigated the German hard-metal industry, led by Mr. Trapp, and the condensed version of which was referred to this morning, is to be issued as B.I.O.S. 1385. The report of the team on the industrial diamond industry of Germany, of which Mr. Trapp was also the leader, is to be issued as B.I.O.S. 1448. There are two other reports, for which Lt.-Col. Ivory was responsible, one on powdered iron and powdered-iron driving bands, B.I.O.S. 1323, and the other on tungsten carbide and tungsten-free hard-metals, B.I.O.S. 1076. Another very interesting report, B.I.O.S. 925, is due shortly, covering some of the important research work by Dr. Dawihl.

My next point is rather more general and relates to long-term policy. It is clear that there is likely to be an increased use of powdered products and sintered components for armament as well as for industrial applications. In view of our experience in the last war I should like to urge that as soon as possible the question of standardization, which was raised yesterday, should be given very serious attention. If, at this stage, standardization of techniques and products could be agreed upon, much time and money would be saved and a greater output maintained in the unfortunate event of the outbreak of hostilities ; even for peace-time economy such standardization would be very valuable.

Finally, I should like to compliment the Institute on providing such a unique opportunity, through personal contact and the presentation of papers, for obtaining a comprehensive review on the subject of powder metallurgy.

The Ministry, which has continually to keep in touch with all scientific and technical developments, finds a Symposium such as this of great value, and it is hoped that meetings of a similar character on other subjects will be arranged in the future. It is also very much hoped that arising out of this Symposium there will be formed an organized, perhaps one should say " compact," body, representative of all aspects of the powder-metallurgy industry, to which the Ministry and other interested groups can turn for authoritative information and practical collaboration.

Mr. **S. J. Garvin** (Murex, Ltd., Rainham, Essex) : I agree with Dr. Jones as regards the D.P.G. Disc atomizing process. During an investigation into German methods of producing metal powders, we visited a place where a number of these machines had been used, and the engineers in charge claimed that they broke down every other day. I do not know whether the Austrians are having more success or have developed the machine further.

The Germans were developing a new method of producing atomized iron powders which consists of the air atomization of a high-carbon iron, and the resulting particles then have to be reduced. That seems to be a very interesting process, because it produces large quantities of powder comparatively cheaply, and the particles are spherical in shape, soft, and can be made to retain part of their carbon, or the carbon can be eliminated completely by reducing in hydrogen.

On the two papers relating to iron–copper compounds, or iron–copper parts made by sintering, it

struck me that nobody has referred to the precipitation-hardening properties of the iron–copper phase system. The diagram shows that the solubility of copper in iron decreases from the order of 8% to somewhere below 2% between 1050° C. and room temperature. Most of the sintering is done at temperatures above 1050° C. to make the best use of the addition of copper, as Dr. Northcott points out. Part of the copper must therefore go into solution in the iron during the sintering treatment and be precipitated during cooling. Conditions favourable to precipitation hardening do therefore exist, and the mechanical properties of the compacts will depend on the rate of cooling. This appears to be borne out also by the test results given in the paper by Northeott and Leadbeater, where Fig. 1 on page 143 shows the influence of copper content on the properties of sintered compacts. There they obtain the maximum tensile properties with a copper content of approximately 24%. In Fig. 16, on page 150, they show the relationship between tensile strength and copper contents for different porosities, and in each case the maximum tensile is obtained for about 10% of copper. I do not know whether it is the case, but I am wondering whether the specimens were cooled under differing conditions and the variation of the cooling rates has had anything to do with this apparent discrepancy in the results.

Mr. **C. E. Richards** (Post Office Research Station) : Dr. Desch raised a question which keeps on cropping up, about carbonyl iron. Carbonyl iron, when first made by the decomposition of iron carbonyl, usually contains about 0·6–1% of combined carbon, and about 0·25% of nitrogen, and in those circumstances it is exceedingly hard. That is the material which is used for most magnetic purposes. The carbonyl iron which is described in most of the textbooks, and which is used for powder metallurgy generally, has been decarburized, usually by the hydrogen treatment; it is down to about 0·1% of carbon, and is completely soft. The Americans have an additional decarburized grade which I understand—I do not know this from personal observation—is largely used. It is known as grade *L*, and is not available in this country in quantity. It is not even spherical; it has been through some treatment which produces a non-spherical carbonyl-base material. It has been advertised as an ideal material for powder metallurgy.

Mr. **J. W. Lennox** (Sintered Products, Ltd.) : While we are talking about carbonyl powder, it could be said that one of the important factors in using powder is the price, and today the price of carbonyl powder is in the region of 4–7*s.*/lb. The type of powder which we as fabricators of powdered parts are interested in must be around 6*d.*/lb.

Yesterday we were talking about onions, and it caused some amusement. One of the reasons why carbonyl powder is not used, apart from the price, is the fact that it is extremely difficult to press, even in the decarburized state. I think that the so-called onion structure has much to do with the poor pressing properties. Excessive cracking is found in the compacts. We have spent some time in trying to get reasonably good carbonyl-iron pressings, but in the end we have had to resort to a method of presintering and then pressing to provide pure iron for special work.

Referring to the paper by Mr. Burr and Mr. Clarke, I think that one very important piece of information has been omitted, in that the authors have not stated the method of pressing. For me, the speed at which the part is pressed is important, and also the type of die finish. With regard to Dr. Jones' remarks about lubricating the die by a mechanical method, it will be seen that these experiments were carried out with a die lubricated with a thin film of oil and graphite. During the pressing of the powder, which would be appreciably longer than the finished compact, the powder tends to break through the so-called lubricating film and die attack results. I think that if this paper had contained a simple diagram of the die and of the method by which the pressure was applied, it would have been far more complete.

Lubricating the powder in place of the die wall is quite sound commercially. The work of Wulff and others,* as discussed by Dr. Jones, shows what happens to the powder during pressing. A powder mixture containing a solid lubricant tends to be warm during pressing and a continuous film is maintained at the die wall during the pressing operation. The inclusion of stearin, which the authors mention as appearing to have detrimental effects on the physical properties, is interesting, but I think that the authors should give some description of how this was added to the powder to obtain uniform distribution. In later papers reference is made to 0·1% of solid lubricant, and to improved results being obtained with this small percentage. We have found that quantities of the order of 0·1% are quite sufficient, provided that the necessary care is taken. I think it is Dr. Maurice Cook who refers to this matter.

Referring to the paper by Mr. Judd, on page 118 he states that pressures of 100–150 tons/sq. in. may be used, provided that such high pressures are employed only for small components of simple shape. From the experience which we have had to date with the pressing of metal powders, I think that when you get past 40–50 tons/sq. in. you are asking a good deal of the die-steel manufacturer. I do not know whether any of the steelmakers will let us have a steel which can be used in compressing metal powders to high densities with pressures of 100–150 tons/sq. in., but if there is such a steel we should very much like to have it. Mr. Judd also states that he manufactures piston rings. I think that this is not quite true, and that it is a piston-ring material which is made rather than a piston ring. One might visualize a piston ring turned out of a die and just put through the furnace, but I do not think that that is what is happening; it is only the material which is being made, probably to a simple ring shape.

On page 122 Mr. Judd states that nickel can be

* R. Kamm, M. Stienberg, and J. Wulff, *American Institute of Mining and Metallurgical Engineers, Metals Technology*, 1947, February, Technical Publication No. 2133

fairly readily diffused at temperatures of the order of 1100° C. in a time of 30 min. We have carried out fairly extensive work with many iron powders with the inclusion of nickel but we have not yet been able to get the nickel to diffuse ; we can always find the nickel afterwards as nickel.

Most of the papers have stated the type of powder, but as a powder user I feel that it would assist us greatly if the many authors would, instead of referring to powders by numbers, refer to the origin of the powder, because in many instances we should like to get powders of the type described, but from the general description given in the text, we cannot do so.

Mr. **C. E. Richards** (Post Office Research Station) : With regard to Mr. Lennox's remarks that part of the difficulty of pressing carbonyl powder may be due to the onion structure, I would say that when carbonyl " onions " are cooked, the onion structure disappears far more completely than when the ordinary domestic onion is cooked—there is no trace of it left ; there is just solid high-purity iron. The powder used for powder metallurgy is soft and has retained no trace of its original onion-skin structure.

Mr. **Gordon Smith** (The Mond Nickel Co., Ltd., London) : Dr. Desch has referred to the use of carbonyl-iron powder in America, and Mr. Lennox hit the nail on the head when he said that it was the high cost of this powder which prohibited its employment for such uses as we are discussing in this section. Mr. Lennox also referred to the difficulties in pressing. I do not think that such difficulties have been confirmed in other directions. In Dr. Pfeil's paper there was a reference to American and German work, and this powder in the decarburized state to which Mr. Richards has referred is quite satisfactory for pressing. It is possible that in Mr. Lennox's experiments there may have been some superficial oxidation, as the powder is very fine and is subject to oxidation if not properly stored ; but there may be other reasons for the failure.

The point is that this powder is a very useful tool for research in this field because of its high purity. I think that is why the Americans have been using it, and it was used in Germany in most of the early stages of German developments in iron powder metallurgy ; they started off with carbonyl-iron powder, and when going into production found what necessary modifications had to be made to make a cheaper powder.

Dr. **L. Northcott** (Armament Research Department) : I should like to raise a question on Chadwick, Broadfield, and Pugh's paper on iron–copper compacts. On page 153 they say : " Hardness and strength increase progressively with increasing copper content, and reach a maximum with 20–25% of copper . . . . The improvement in mechanical properties is, however, largely accounted for by mutual diffusion, the solid solutions being of higher strength than the pure metals." I should like to ask the authors whether an annealing treatment of ¼ hr.—I am prepared to assume 4 or 5 hr., but I think that ¼ hr. is what they use—is going to diffuse or permit diffusion of copper into iron or iron into copper. The iron will certainly diffuse into the copper, partially at any rate, because the copper is in the liquid state ; but I cannot imagine much diffusion of copper into solid iron.

The same point arises with regard to nickel. Apart from carbon and hydrogen and one or two other elements, we know from general steel practice that the solid metals do not readily diffuse until a very high temperature, a little below the melting point, is reached and kept there for a considerable time.

My own view is that the improvement in properties, the increase in strength, of the iron–copper or iron alloys with the presence of copper is due to the extent of 99% to the increased bonding action. The question of cohesion has already been raised this afternoon. The strength of sintered components of the engineering type is largely due to the cohesion between the different individual particles, and any treatment such as the existence of a liquid phase, or, if there is no liquid phase, any treatment which will render the surfaces more readily cohesive to one another, will increase the mechanical properties.

Mr. Garvin raised the point of precipitation hardening. We have shown that the precipitation of iron in the copper phase does occur, but any increase in properties due to such a precipitation-hardening effect is, I feel convinced, very small—I myself would call it negligible—compared with the benefit to be obtained from the increased cohesion between one particle and another owing to the presence of the liquid phase.

Mr. Garvin referred to Fig. 16 in the paper by Mr. Leadbeater and myself, where the peak in the tensile-strength/copper-content curve occurs at about 10%. We have attempted to explain that, but what we say is speculation, and I should be glad to hear any other suggestions during the course of the meeting. We have attempted to explain that with 10% of copper under the conditions which we use we have the maximum penetration per unit weight of copper and the formation of the optimum thickness of copper as bond.

Mr. Lennox raised a very interesting point on the question of the origin of the powder. My experience is that manufacturers will supply a powder to any particular specification, but powder metallurgy being such a new science we often do not know what we want, and at the moment we are trying to find out. If we say that powder No. 29 was not nearly so successful as No. 32, and if No. 29 came from manufacturer *A* and No. 32 from *B*, there would be complaints from manufacturer *A*.

We have tried carbonyl iron. Apart from the question of expense, which is easily the most important single factor, it is not a convenient material to handle, largely, I think, owing to its fine particle size. It does not flow readily, and in our experiments, which were deseribed yesterday, on flowing properties, not only with carbonyl iron but with certain other iron powders, we have noted the resistance to flow of certain powders. It is amazing to see a cone full of powder with a cone orifice of 14 mm. in dia., and the powder refusing to flow through that orifice.

Lt.-Col. **W. Ivory** (Armament Research Department, Fort Halstead, Sevenoaks, Kent) : The Germans

used a huge tonnage of powder and committed themselves to the sintered-iron driving band to the exclusion of all other materials. In 1940 they were in serious trouble owing to the indiscriminate use of all kinds of powders without any regard to their physical characteristics. Extensive research on methods of manufacture of the band failed to effect any improvement, and it was not until they studied the physical properties and grain characteristics of the various powders that they met with any success. They then determined suitable powder mixtures for driving bands, fuse bodies, etc., and specified them to the manufacturer. It would appear that the British manufacturer should determine the class of powder suitable for his particular application and not select any powder that may be offered.

With reference to the strength of sintered powder components, I came across an interesting application which has not been mentioned. The Germans were sintering ball-bearings and using them. Dawihl stated that they were reasonably satisfactory.

AUTHORS' REPLIES

Mr. **C. J. Leadbeater** wrote in reply: Dr. Jones's remarks (*see* page 34) are more pertinent to the Hametag process, which had been mainly involved in the production of the large output of iron powder rather than to the D.P.G. atomization process, which had been developed to a smaller degree as borne out by the data in Table I, p. 193. Despite the greater rate of output of the Disc process, it was disliked on account of personal danger and high maintenance; it is estimated that the number of machines installed throughout Germany did not exceed 6–12.

Mr. **R. Chadwick**, Mr. **E. R. Broadfield**, and Mr. **S. F. Pugh** wrote in reply: Dr. Jones has criticized the value of the D.P.G. process on the grounds that the powder is in an oxidized condition, and needs to be reduced by heating in hydrogen before it can be used. We would like to point out that this is true for most other methods of powder manufacture. The product has more or less surface oxidation, and a reducing treatment is therefore necessary. The rather high proportion of oxide in the D.P.G. powder scarcely affects the cost of the annealing operation.

Dr. Northcott suggests that the high strength of the iron–copper compacts is accounted for by the bonding action of the copper, and doubts whether diffusion of copper into the iron particles would be sufficiently rapid to be appreciable at the temperature of sintering. In reply we would like to point out that in sintering, the melting point of one of the constituents is exceeded. The extent of reaction can be seen from the rounding of the iron particles. If, for example, Fig. 15 in our paper on iron–graphite is compared with Figs. 7 and 10 in the paper on iron–copper alloys, it will be seen that the change in structure cannot be wholly explained by the solution of the surface layers of the iron particles by molten copper.

We have attempted to obtain some evidence on the solid solution of copper in the iron-rich phase by the determination of lattice parameters. Diffraction patterns on sintered and slowly cooled iron-rich alloys showed diffuse lines on which no measurement of parameter was possible, and study of the iron–copper equilibrium diagram* indicates that this might arise from strains resulting from transformations occurring at 833° C. during cooling. In order to avoid such strains, iron–copper compacts sintered at 1200° C. were held at 800° C. for a period of 24 hr., after which they were alternatively either quenched or slowly cooled to room temperature. After such treatment the pressings gave sharp diffraction lines and it was possible to make lattice-parameter measurements within an accuracy of $\pm$ 0·0002 kX. The following figures show the values obtained:

	Parameters of Sintered Compacts, kX.	
	Quenched	Slow-Cooled
100% Iron	2·8604	2·8611
3% Copper	2·8631	2·8620
5% Copper	2·8631	2·8616
10% Copper	2·8633	2·8635
25% Copper	2·8626	2·8617

The accepted value† for the lattice parameter of pure electrolytic iron is about 2·8604 kX. at room temperature. According to Norton,‡ the increase in lattice parameter for solid solution of 1% of copper in the iron lattice is about 0·0014. The above data would appear to indicate that copper of the order of 2% is, in fact, retained in solution in the compacts after prolonged treatment at 800° C., while in slow cooling a variable proportion is precipitated, the sharpness of the diffraction lines on the whole series of samples indicating the substantially uniform solution throughout the iron-rich grains. There would not appear to be any easy method of determining parameters from material under the conditions obtaining at the sintering temperature, but it would seem likely that substantially more than 2% is present.

In the light of this evidence Dr. Northcott's theory must be partially rejected, and it is clear in considering a structure such as that illustrated in Fig. 7, that if the iron-rich grains are substantially weaker than the cement, fracture could occur almost entirely through the iron grains which, if unalloyed, would have no more strength than pure iron. Whether the solid solution of copper in the iron can fully account for the increased strength, it is not possible to say, and it may be that a complex cellular structure has mechanical properties which cannot be simply explained in terms of those of the two separate constituents.

We would agree with Mr. Leadbeater that the plot shown in Fig. 2 does not in effect show a true relationship between mechanical properties and copper content, since porosity and copper content are changing at the same time. We have attempted to replot our data in a similar manner to that shown in Fig. 16

* M. Hansen, "Aufbau der Zweistofflegierungen," p. 560. Berlin, 1936: Julius Springer.

† A. Taylor, "Introduction to X-Ray Metallography", p. 383. London, 1945: Chapman and Hall.

‡ J. T. Norton, *American Institute of Mining and Metallurgical Engineers, Metals Technology*, 1934, Dec., Technical Publication No. 586.

of the paper by Northcott and Leadbeater, and while we would prefer not to reproduce these curves, since they are based on extrapolations which may not give accurate porosity values, the results seem to indicate that the proportion of copper giving the true maximum strength is somewhere in the region of 10–15%, which is a substantial confirmation of these authors' findings.

Part 2—Non-Ferrous Components (Based on Papers 23 to 27 of the Symposium) *

Dr. **Maurice Cook** (Imperial Chemical Industries, Ltd., Metals Division, Birmingham) (rapporteur): Mr. Tait's paper is a concise and interesting account of powder-metallurgy methods used in this country for the manufacture of plain bearings and thrust washers, in which he first of all discusses the qualities of these methods which make them so valuable, for with their aid it is possible to produce porous structures which will hold oil, and also alloys, metallic mixtures, and structures difficult or impossible to produce by other processes.

The idea of porous bearings originated in the U.S.A. as far back as 1916, and since that time the development in that direction has been very considerable, until now the output of porous bearings is actually reckoned in millions per annum. In this country there have been no particular or outstanding developments in this branch of powder metallurgy, and production is generally along American lines.

The author goes on to describe the technique usually employed for the production of porous bronze bearings, and the materials used—90/10 copper–tin bronze containing up to 1·5% of graphite, copper–tin–lead alloys to a lesser extent, and also iron–copper alloys, which are not used very much in this country. They have a higher load-carrying capacity at slow speeds than the tin and tin–lead bronzes, but at moderate and higher speeds they are inferior to the bronzes.

Porous materials are not so strong as the corresponding solid metal, while the frictional heat generated is not dissipated by a flow of oil as it is in normal lubrication practice, and must be lost by conductivity through bush and housing, or journal. Furthermore, the discontinuous nature of the surface hinders the formation of a full-fluid oil film. In general, therefore, porous oil-retaining bronze bushes are limited to a PV value of about 50,000, where P is the pressure in pounds per square inch of projected bearing area, and V the peripheral velocity of the journal in feet per minute. They can seldom be used at velocities exceeding 500 ft./min., because at higher speeds the rate of heat generation is likely to exceed that of heat dissipation.

In the field of non-porous bearings, attention has been chiefly directed to copper–lead and lead–bronze materials, bonded to a steel back. The procedure is to spread the mixed metal powders loosely to a uniform thickness on the steel strip and sinter in a reducing atmosphere. The sponge produced in this way is compacted by pressing or rolling until the voids have substantially disappeared. The strip is finally sintered, and bearings, bushes, and the like are produced by normal pressing methods.

Among newer developments not yet commercially available, the author refers to steel-backed porous bronze bearings, and also to oil-well porous bushes, *i.e.*, a composite oil well and porous bush.

The first of two appendices to the paper is concerned with the influence of structure on the behaviour of the copper–lead bearings. Two types of structure can be produced by casting, but the third type, which is referred to as two interlinked sponges, and which inferentially is the best, can be produced readily and consistently by carefully controlled powder-metallurgy methods. The second appendix deals with the advantages in bearings of the low elastic modulus associated with porous metallic materials.

The paper by Mr. Pugh and myself deals with an investigation into the pressing and sintering of electrolytic, oxide-reduced, and chemically precipitated copper powders of various particle-size gradings. The methods used for evaluating these powders, for compacting the ring products and testing them are described in some detail. Following a number of preliminary experiments, the incorporation of 0·1% of zinc stearate was adopted throughout in all the experimental pressing. Observations have been made on the compressibility of the various powders, and the proportionality of tensile strength to ductility in sintered copper, has been confirmed. A comparison was made of the pressing and sintering of one size grade, M, of the three types of powder, with pressing loads ranging from 15 to 65 tons/sq. in. and sintering temperatures from 750° to 1050° C. The results of these experiments are shown in Figs. 6, 7, and 8, and the dimensional changes and tensile strength of compacts made from general grades of electrolytic and reduced powders, sintered at 1020° C., are illustrated in Figs. 29 and 30.

Three series of photographs are included, which show for each of the three types of copper the external appearance of the powder particles, their internal structure, the structures of unsintered compacts, and the structural changes resulting from sintering at progressively increasing temperatures. It is of interest to note that the electrolytic and the precipitated powders have microstructures similar to that of wrought copper, while the reduced powder showed no visible grain structure even at very high magnifications, but numerous fine dark markings, which appeared to be minute fissures, possibly resulting from the processes of oxidation and subsequent reduction. This spongy structure might be expected to have a

* The following corrections should be noted to the first edition of the Symposium:

On page 171 the first sentence in the right-hand column should read: "Unsintered compacts pressed from hydrogen-reduced powder, Fig. 17, show no visible change in appearance on sintering at 400° C., Fig. 18, ..."

On page 172, lines 12, 15, and 18, for "precipitated" read "reduced."

On page 187 the inscriptions to Figs. 6 and 7 should be transposed.

much larger capacity for absorbing oxygen and other gases, and may even retain oxides in these internal fissures after low-temperature reduction in hydrogen. The evolution of adsorbed gases during sintering, and the reaction with internally adsorbed or combined oxygen, would offer an explanation of the general tendency shown by these results for expansion and low tensile strength in compacts prepared from oxide-reduced powders.

Electrolytic and precipitated powders can be pressed under moderate loads to yield, after sintering, products of relatively high tensile strength, *i.e.*, of the order of 10–12 tons/sq. in., with small dimensional change, whereas reduced powders when pressed under conditions giving maximum strength show large dimensional changes which vary substantially with relatively slight changes in sintering conditions.

Mr. Lennox describes experiments carried out with the object of establishing optimum techniques for the production of high-density components in 70/30 brass and 90/10 tin bronze, with maximum tensile strength and elongation values. Products were made in the form of $\frac{1}{2}$-in. and $1\frac{1}{2}$-in. dia. pellets on 30-ton and 200-ton single-acting hydraulic presses, and thereafter sintered in hydrogen. The 70/30 pellets were produced from mixed copper and zinc powders, from mixtures partially alloyed by prior heating at 450–550° C., and from completely alloyed brass powder. Compacts prepared from unalloyed powder mixtures expanded considerably with evolution of zinc vapour, and distortion was pronounced, especially at the highest sintering temperature employed, 925° C., which is very close to the liquidus temperature for this alloy. Partially alloying the mixed powders before pressing resulted in considerable improvement, but the best results were obtained with the fully alloyed powder. The maximum density was obtained with the highest pressure used, 50 tons/sq. in., on sintering at 810° and 900° C.

The effect of re-pressing and re-sintering was also investigated. The economic and other limiting factors in using high pressures for cold-compacting were considered, and led to the determination of the physical properties which could be expected under commercially practicable conditions. As a result of these tests, the conclusion was reached that the preferred procedure for making the 70/30 brass alloy was to use alloy powder with a die lubricant, cold-pressed at 30–35 tons/sq. in., sintered at 900° C. in hydrogen or cracked ammonia, followed by re-pressing and re-sintering as before. It is considered that the reduced die wear and the increased accuracy in dimensions obtained by resorting to re-pressing and re-sintering warrants the extra cost of this technique.

In contrast with brass products, bronze compacts were satisfactorily prepared from a mixture of the constituent metal powders, observations being made on the properties of compacts pressed at loads varying from 10 to 30 tons/sq. in., with sintering temperatures of 800° and 850° C. Less satisfactory results were obtained with partially alloyed copper and tin powders, while with atomized alloy powders it was not always possible to obtain compaction.

Very fine powders were found to yield the best properties, presumably because of the rapid inter-diffusion of the constituent metals, but such fine powders do not flow readily and are therefore not suitable for commercial production unless used in conjunction with a vibratory mechanism.

Various methods of adding tin were also investigated, comprising additions of alloy powder in different proportions to copper powder or to a copper–tin mixture. Flow properties were usually improved, at the expense however of high sintered density. For fine copper and tin mixtures the conditions likely to produce the best physical properties are indicated as being cold-pressing at 30 tons/sq. in., sintering at 800° C. in hydrogen for 30 min., re-pressing at 30 tons/sq. in., and sintering as before.

Finally in this paper a number of brass and bronze components are illustrated and described, and various aspects of the mass production of such components with reference to die and press design are discussed.

After considering briefly the particular features associated with the pressing of aluminium powder arising from the presence of oxide, Mr. Bickerdike describes experiments on the pressing and sintering of mixtures of aluminium powder with 6% of copper powder. Discs measuring $1\frac{1}{8}$ in. in dia. and $\frac{1}{4}$ in. thick were prepared under a variety of conditions of hot- and cold-pressing and sintering in air, vacuum, and hydrogen, and density, hardness, and tensile-strength values are given, with a series of photomicrographs.

With a sintering temperature below the eutectic temperature of this binary series, the density increased with the compacting pressure at a gradually decreasing rate until the compact pressed at 50 tons/sq. in. contained only very slight porosity after sintering. When sintering was carried out above the eutectic temperature, the density reached a maximum at a given value of the cold-pressure and decreased after the pressure was raised beyond this.

The cold-pressure and sintering temperature have a critical effect on the properties of the sintered material. In some of the pressure/sintering-temperature range considered, an alteration in the pressure of 5 tons/sq. in. can cause a considerable change in tensile strength, while an increase of 15° C. in sintering temperature can produce a change from a dense material to a highly porous one. Different combinations of pressure and sintering temperature can lead to similar tensile properties but considerable differences in microstructure. The advantage of sintering below the eutectic temperature is that the surface details of the specimen remain unchanged, whereas that is not so when partial melting occurs.

When the sintering of solid phases is relied on to produce material with high mechanical strength, fairly high compacting pressures are necessary to promote diffusion and recrystallization on heating. When material is sintered at a temperature at which it is partly molten, greatly accelerated diffusion occurs and a characteristic distribution of voids is brought into existence, which diminishes by shrinkage after long periods of heating. Tensile-strength and elongation values of 21·3 tons/sq. in. and 4·5% respectively have been obtained with compacting pressures of

20 tons/sq. in. and a sintering temperature of 575° C., indicating that, provided surface details are not of great importance, it is not necessary to use more than moderate compacting pressures in order to obtain reasonably good tensile properties with small specimens.

Mr. Leadbeater's paper shows that in non-ferrous metal powders the major interest in Germany, as measured by tonnage, was in aluminium powder, which in flake form achieved an output of 13,000 tons per annum. It was obtained by atomizing or disintegration of foil and afterwards pulverizing in heavy, medium, and fine stamping machines.

The equipment used in general is similar to that used in this country but various improvements in design have been made. Frequent screening at appropriate stages during stamping improved the quality of the powder by eliminating over-stamping. One factory manufactured flake aluminium powder by ball-milling, and a satisfactory product was obtained by using paraffin wax instead of stearin as lubricant.

There is a brief reference to the manufacture of copper powder by electrolytic methods.

Most of the flake aluminium powder produced in Germany was used in such applications as torpedo-head charges, land and sea mines, flares, and for the chemical industry, but reference is made to the use of a 90% iron–10% aluminium mixture for the manufacture of heating elements, to cooking plates made from a mixture of 80% aluminium powder and 20% alumina, and to the de-icing parts made of copper in tubular form with a 70% porosity, impregnated with ammonium chloride.

Mr. **J. W. Lennox** (Sintered Products, Ltd.), opening the discussion, said: The paper by Dr. Maurice Cook and Mr. Pugh shows a very good die design for experimental work and I should be particularly interested to know the clearance between punch and die, because on page 164 it is stated that all the flashes were removed from the compacts after pressing. This to me indicates that there was quite a reasonable clearance in the die during compression.

Standardization has been referred to a number of times, and all the work has shown that slight variations in powder will give the fabricators of parts a great deal of trouble. I would emphasize, as I have before, that the papers should give the origin of the powder, and that the various grades should be standardized and made such that both user and manufacturer can check on each powder before using it for production purposes.

Mr. **R. Chadwick** (Imperial Chemical Industries, Ltd., Metals Division, Birmingham): Mr. Bickerdike's paper is extremely interesting and several of the more important points which arise in the powder metallurgy of aluminium alloys have been touched upon, but I think the author will agree that his investigation does little more than point out the possibilities while, at the same time, it raises a number of problems on such points as the effect of oxide, the effect of gas absorption, and the temper-hardening reactions in the sintered product.

The author points out the importance of low oxygen content, but he has omitted to determine oxygen on his powder sample, which I think is a serious omission, for another powder sample of differing oxygen content might give substantially different results. Our own work on the pressing of aluminium powders has certainly suggested that the oxygen content has a very great effect on sintered strength, and that from commercial powders containing up to 1% of Al_2O_3, it is necessary to select a powder with less than 0·1% for powder-pressing.

Mechanical properties are quoted only for compacts quenched and fully aged at room temperature, and it would have been interesting to know how much of the strength resulted from age-hardening. It seems unlikely that some of the mechanical properties quoted could have been obtained without substantial age-hardening, but on the other hand, according to available published information on the temper-hardening of aluminium–copper alloys,* the presence of 0·28% of iron would be likely to reduce the age-hardening of the alloy to an almost insignificant amount. In our investigations we obtained hardness values after quenching and ageing of about the same order as the present author on pressed and sintered aluminium–copper alloys, and the results appeared to be at variance with Petrov's findings; it may be, therefore, that the conclusions reached for wrought materials are not entirely valid for powder-pressed products. On the other hand Petrov's conclusions were not based on any single piece of direct experimental evaluation, and there would seem to be a certain amount of doubt about the validity of his conclusions in respect of the straight aluminium–copper alloy.

The only really satisfactory way of following the diffusion process in an investigation of this character is by the X-ray measurement of lattice parameters. In our work on the sintering of a 95/5 aluminium–copper powder mixture, we found parameters corresponding to a solution of from 2 to 4 in. of copper after 2 hr. at 580° C. There would seem therefore to be a reasonable possibility of substantially completing diffusion within a short period, and the long sintering periods employed by the author would not appear to be justified.

Mr. **R. L. Bickerdike** (Royal Aircraft Establishment, Farnborough): No age-hardening experiments were done on the 6% copper–aluminium alloy; in all cases a period of three weeks at room temperature elapsed before mechanical tests were carried out on the quenched specimens. The oxide content of the aluminium powder used was not determined.

It seems likely that hydrogen evolved during cooling from the higher sintering temperatures contributed to the formation of the large pores observed in those specimens which had initially been pressed to a high density. No experiments were made, however, to confirm the origin of the gas entrapped within these compacts.

Mr. **P. D. Liddiard** (Glacier Metal Co., Ltd., Alperton): I must apologize for Mr. Tait not being

* D. A. Petrov, *Journal of the Institute of Metals*, 1938, vol. 62, p. 63.

here, but unfortunately he is indisposed. Dr. Maurice Cook appears not to like references to vegetables. Apparently he did not mind Mr. Tait's reference to a mixed copper and lead sponge, but no doubt that is because the porifera are members of the animal kingdom ! I think that Dr. Cook would have been offended by one of Mr. Tait's favourite descriptions of one of the cast types of alloy. Mr. Tait mentions the two types, one in which there is a dendritic structure of the copper with a continuous lead phase, and the other in which there are discrete lead particles in a continuous copper phase ; this latter he usually describes as " currants in a bun," the currants being a vegetable phase ; I do not know whether one would call a bun nowadays a mineral phase or not.

I am sorry that there has not been more discussion on this section of the Symposium. Mr. Tait's paper brings out one of the weapons of powder metallurgy which has not been sufficiently emphasized before, namely, its use where there are immiscible materials, materials which normally could not be obtained in a suitable condition by cast methods. If anyone here has had experience with the casting of copper–lead alloys, he will know that it is extremely difficult to control the size and distribution of the lead. Powder metallurgy has provided a method for doing that. The network so obtained, a continuous network of lead, does ensure lead coming to the surface during running where it is required for lubrication. It also allows for a continuous phase of copper, to give the necessary strength. These materials have a steel backing, but they also need a continuous copper phase to give fatigue resistance.

Some of the earlier efforts in this direction were unsatisfactory because of residual porosity, and we are dealing here with a bearing which does not require porosity. The bearing as produced today has no porosity at all ; it is a fully compacted material of suitable structure.

The diagrams in the first appendix to Mr. Tait's paper are taken from work by Dr. Bowden and his co-workers at Cambridge. I was asked this morning whether this was scientific fact or sales propaganda. I do not think that Dr. Bowden has any sales axe to grind, and I should certainly recommend all those who are interested to read his work on this subject, and particularly the Australian paper, which is mentioned in a footnote on page 160, which deals with work largely done in Australia during the war, and now being continued at Cambridge.

I have one comment on the paper by Dr. Cook and Mr. Pugh. It would have been much more complete if it had been possible to include an examination of the spherical powders as well. I do not think that spherical copper powder is readily available, but if small quantities of it could have been made and examined it would have made the paper more complete.

In the previous section of this Symposium reference was made to ball bearings being made by powder-metallurgy methods. I am wondering whether the speaker who made that remark was under the impression that ball-bearing substitutes were ball bearings. The Germans were making ball-bearing substitutes which contained no ball bearings, but one ring, usually of sintered iron, in a ring of sintered steel.

AUTHORS' REPLIES

Dr. **Maurice Cook** : In reply to Mr. Lennox, the clearance between punch and die in the apparatus for making the ring compacts used in our investigations was 0·001 in. In evaluating the quality or suitability of a powder it would not seem that knowledge of its actual source would necessarily be of much help, but rather information regarding the method or process used in its manufacture, and adequate details of its physical and chemical properties and characteristics which, in turn, emphasizes the need, already mentioned several times in this Symposium, for standardized procedures for examining and testing metal powders.

Mr. **R. L. Bickerdike** wrote in reply : As Mr. Chadwick has remarked, there are several important aspects of the sintering of aluminium alloys not mentioned in my paper. It is not known how much age-hardening, if any, occurred during the three-week period at room temperature which elapsed between quenching and mechanical testing.

The alumina content of the hydrogen-sintered compacts for which mechanical test results have been quoted was found to be <0·4%. It is possible to sinter powders with a considerable content of alumina when the latter exists mainly in discrete aggregates and not as a uniform film.

Further Author's Reply to the Discussion on SECTION D—HARD-METAL CARBIDES

Mr. **G. S. Trapp** wrote in reply : The limitation imposed on the length of the paper placed a severe restriction on the subject matter, and discussion of ammunition-core production was omitted as other developments were thought to be of more general interest. The total German output of cores was over 2500 tons and not 1500 tons as stated by Mr. Chadwick. Furthermore, use of the hot-pressing process did not eliminate grinding of the ogive (the radius at the tip) ; in fact it increased the difficulty of grinding the ogive since the method of hot-pressing which was adopted left a shoulder at the point at which the ogive joined the parallel diameter.

The incorporation of sintered scrap in production of new cores by hot-pressing was tried experimentally in the United Kingdom during the war. It was considered that the imperfect bond between the small core and the surrounding material would weaken the structure and possibly leave stresses, and that these defects would increase the tendency of the core to

shatter on impact with the target instead of penetrating it. The acute shortage of tungsten in Germany compelled the adoption of any such device which would increase core supplies without increasing the demand for tungsten.

It has never been suggested that the extensive Krupp hot-pressing plant was put down for the production of substitute carbides. Nevertheless, it was used for a number of purposes apart from ammunition cores, and an account of all the applications, including core production, is given in B.I.O.S. Final Report No. 1385.

Owing to the limited length of the paper a full description of the methods employed and the results of determining the characteristics of carbides by measurement of electrical conductivity and magnetic properties could not be given. These also are given in greater detail in the B.I.O.S. report.

Correspondence on the Paper—

ABNORMAL CREEP IN CARBON STEELS* *by J. Glen*

Mr. **W. B. Brooks** (Alloys Development Co., Pittsburgh, Pa., U.S.A.) wrote: Mr. Glen's paper is a distinct contribution to the literature on creep. It is another step in resolving the various aspects of creep as related to deoxidation practice, grain-size, heat-treatment, etc. Mr. Glen notes the significance of ageing phenomena and the fact that creep rates cannot be wholly resolved in terms of grain-size.

The writer would submit that further light can be shed on the effect of aluminium by examination of its effects on ageing. If various alloying elements are arranged in order of increasing effectiveness as solution hardeners and creep retarders, some suggestive differences may be noted. Silicon, for example, is a potent solution hardener but a rather poor creep retarder. The greater effectiveness of the carbide-forming elements as creep retarders, even in low-carbon irons, seems significant. More significant is the effect of heat-treatment on an aluminium deoxidized steel noted by Cross and Simmons.† Air-cooling after heating for one hour at 2000° F. gave a creep rate of 0·00006 in./in./hr. under the conditions of test, while furnace-cooling or subsequent normalizing at 1730° F. gave rates of 0·016 and 0·047 respectively. In the case of the air- and furnace-cooled specimens, the austenitic grain-size variable was eliminated. One might take this to mean that some trace element is kept in solution for subsequent precipitation during creep in the air-cooled specimen, and that it is wholly precipitated by furnace-cooling or normalizing.

The kinetics might be somewhat as follows: Thermal oscillation of the lattice components at creep temperatures permits the generation and translation of dislocations on a large scale even in the presence of solution hardeners. The highly strained state of the lattice surrounding a Guinnier-Preston particle does not permit the generation and diffusion of dislocations on such a large scale at creep temperatures. Presumably, the Guinnier-Preston particles are carbides and nitrides, and aluminium has a two-fold effect: (1) It affects the distribution of carbon in several ways and promotes graphitization and the coalescence of carbide particles; (2) it combines with nitrogen on slow cooling or normalizing, thus preventing the subsequent formation of nitride Guinnier-Preston particles during creep. This hypothesis lends emphasis to the author's caution on the utility of short-time tests.

Mr. **W. E. Bardgett** (The United Steel Companies, Ltd., Sheffield) wrote: The deleterious effect on creep resistance of aluminium and the beneficial effect of silicon added in certain proportions to mild steel have been known for some considerable time, but the systematic work of the author raises appreciably the level of our knowledge of this subject from a quantitative aspect. The results refer to tests carried out under particular conditions of stress, temperature, and time and it is to be borne in mind that any departure from these conditions may lead to quite marked differences in the relative effects of the factors under consideration.

In the tests carried out by the author, comparisons were made on the basis of creep rate and as these cannot, from the data given, be stated in terms of stress, it is of interest to compare results of tests carried out in the Central Research Laboratories of The United Steel Companies, Ltd. on this basis.

The method of test adopted on the experimental steels was that known as the Barr-Bardgett accelerated test.‡ This depends on the measurement of the diminishing rate of creep in a stressed test-piece by means of a steel weigh-bar, so that the creep of the test-piece is accompanied by a proportionate decrease in the applied stress. Comparative tests have shown that results obtained by this method in terms of stress are similar to those obtained by long-time tests at stresses corresponding to small rates of creep of the order of 10^{-8} in./in./hr. in the temperature range under consideration.

† American Society for Testing Materials, 1945, Report of Joint Research Committee on "Effect of Variables on Creep Resistance of Steel," Appendix I.

* *Journal of the Iron and Steel Institute*, 1947, vol. 155, Apr., pp. 501–512.

‡ W. Barr and W. E. Bardgett, *Proceedings of the Institution of Mechanical Engineers*, 1932: N. A. Schaposchnikov, *Zavodskaya Laboratoriya*, 1936, 5.

TABLE A—*Analysis and Properties of Experimental Mild Steels*

Steel No.	Analysis						Additions		Grain-Size (McQuaid-Ehn)	Maximum Stress, tons/sq. in.	Yield Stress, tons/sq. in.	Elongation on $4\sqrt{A}$ %	Reduction of Area, %	Izod Impact Value, ft. lb.	Barr-Bardgett Creep-Stress Value at 450° C., tons/sq. in.
	C, %	Mn, %	Si, %	S, %	P, %	Al, %	Si, %	Al, %*							
Series 1															
1	0·12	0·39	0·015	0·035	0·046	...	Nil	Nil	1–2, mainly 1	26·3	21·0	43·0	70·0	87	2·0
2	0·12	0·48	0·07	0·035	0·043	...	0·08	Nil	1–5, mainly 1	28·3	18·8	38·0	66·0	88	1·5
3	0·14	0·48	0·16	0·040	0·046	...	0·15	Nil	1–5, mainly 2	29·4	20·9	40·0	68·0	87	2·0
4	0·14	0·47	0·24	0·036	0·045	...	0·25	Nil	1–4, mainly 2	30·0	23·6	37·0	66·0	89	2·8
Series 2															
5	0·14	0·46	0·05	0·040	0·040	0·021	Nil	0·025	1–3	28·2	21·1	38·5	66·0	84	1·4
6	0·14	0·41	0·035	0·037	0·041	0·042	Nil	0·050	6–7	27·6	22·7	41·0	71·6	89	1·5
7	0·14	0·46	0·05	0·039	0·047	0·076	Nil	0·100	7	28·3	21·2	41·5	66·0	83	1·2
Series 3															
8	0·14	0·41	0·035	0·037	0·041	0·042	Nil	0·05	6–7	27·6	22·7	41·0	71·6	89	1·5
9	0·14	0·44	0·10	0·035	0·042	0·044	0·10	0·05	7–8	28·4	21·7	40·0	70·0	89	1·7
10	0·15	0·46	0·25	0·038	0·046	0·054	0·25	0·05	7–8	30·7	24·5	38·0	68·0	89	2·1
11	0·15	0·48	0·50	0·040	0·045	0·041	0·50	0·05	8	31·9	24·7	37·0	68·0	89	2·4

* 0·045% = 1 lb./ton

Three series of mild steels were made for the purpose of determining the effect of :

(1) Treating with silicon in the absence of aluminium.

(2) Treating with aluminium without the addition of silicon.

(3) Treating with aluminium in the presence of silicon.

All the steels were made in a 35-kVA. high-frequency furnace of the Ajax Northrup spark-gap type, in sillimanite crucibles, from Armco and white-iron base.

Additions were made in the following order :

(1) 75% ferrosilicon (if required).
(2) 80% manganese, 6% carbon, ferromanganese.
(3) Ferrosulphide.
(4) Ferrophosphorus.
(5) Aluminium (if required).

The steels were cast in the form of square ingots tapering from 2½ to 2¼ in. and weighing 18 lb. Each ingot was forged to ⅝-in. bar, which was normalized at 910° C. Tensile and impact tests at normal temperature, Barr-Bardgett creep tests at 450° C., and determinations of austenitic and McQuaid-Ehn grain-size were carried out on all the steels ; the results are included in Table A.

Series 1—Effect of Treating with Silicon with No Aluminium Additions

Series 1 was composed of a steel to which no silicon was added and three steels to which ferrosilicon was added to give silicon contents of 0·07, 0·16, and 0·24% respectively. With the exception of the silicon contents and the somewhat low manganese content of the non-treated steel, the analyses of the four steels were practically identical. So far as concerns the composition, the steels may be considered solely with respect to the effect of silicon on the creep resistance, with the possible exception that a small allowance may be necessary in assessing the creep-stress value of the non-treated steel, on account of the low manganese content. Since a decrease in manganese from 0·48 to 0·39% reduces the creep resistance, as shown by the author, the non-treated steel would have a slightly higher creep resistance if the manganese content were raised to that of the other steels of the series.

Determinations of McQuaid-Ehn grain-size showed that the non-treated steel was coarse-grained, of grade 1–2, while the other three steels were of variable grain-sizes, the steel containing 0·16% of silicon being, however, mainly grade 1 and the remaining steels mainly grade 2.

The austenitic grain-size of all the steels was determined. Austenitic grain-size refers to the actual size of the austenite grains established in the final heat-treatment before testing. The grain-size measurements were made according to the same scale as was adopted for the carburized specimens and the four steels were found to be fine-grained.

The effect of increasing the silicon content is to raise the maximum-stress value, with the result that the values for the four steels differ and range from 26·3 to 30·0 tons/sq. in. It is a well-known fact, however, that the cold tensile strength bears no relation to the creep-stress value.

The creep-stress values show that the steel containing 0·015% silicon (no ferrosilicon added) is superior to the steel containing 0·07% silicon, *i.e.*, equal to the steel containing 0·16% silicon and inferior

to the steel containing 0·24% silicon. Whether or not a non-treated steel is superior in creep resistance to a silicon-treated steel may therefore depend on the amount of silicon added to the silicon-treated steel. The results indicate that deoxidation with silicon has the effect of lowering the creep resistance, while silicon as an alloying element, at least up to 0·24%, definitely enhances creep resistance in steels of the analyses considered.

The marked difference in the strength at 450° C. of the steels in this series indicates that the deoxidizing and alloying actions of silicon have marked and opposite effects on the creep resistance. The corresponding series of steels in the author's paper show a similar trend, with a minimum in the creep resistance at about 0·04% silicon. The main difference, however, in the two sets of results is the apparently markedly greater effect of silicon in enhancing creep resistance in the writer's tests.

Series 2—Effect of Treating with Aluminium with No Silicon Additions

Three steels were made to test the effect of additions of 0·025, 0·050, and 0·10% of aluminium. The remaining elements were kept approximately constant.

The analyses and test results corresponding with those of the previous series are included in Table A.

As would be expected, the inherent grain-size of the steels was found to differ, the steel having an aluminium addition of 0·025% being coarse, and the steels having aluminium additions of 0·042 and 0·076% being fine-grained of grades 6–7 and 7 respectively. Since the difference in grain-size is a result of the treatment with aluminium, it is impossible, in steel differing only in the aluminium addition, to obtain in one set of samples of the selected analyses a constant McQuaid-Ehn grain-size and a variable aluminium content.

Austenitic grain-size developed on normalizing was also determined and the steels all showed a fine grain. The tensile and impact tests were uniform.

The effect of the addition of 0·025% of aluminium is to lower the creep stress from a value of approximately 1·7 tons/sq. in. (steels 1 and 2) to 1·4 tons/sq. in. Raising the aluminium addition to 0·050% has no appreciable additional effect on the creep resistance, but a further increase to 0·10% results in a definite lowering of the creep-stress value to 1·2 tons/sq. in. The effect of aluminium additions in this series is apparently markedly less than in the author's series, the trend of the results being, however, similar. The apparently greater effect of silicon and lesser effect of aluminium in the above two series of steels, compared with the author's results, indicates a lower creep resistance in the basic composition of these series (steel 1) than in the corresponding steel of the author's series (steel 1, Table II).

It is of interest to note that on the basis of these results it is possible, by variation in the aluminium and silicon additions, to obtain two mild steels of 28 to 30 tons/sq. in. tensile strength, one of which has twice the creep strength of the other.

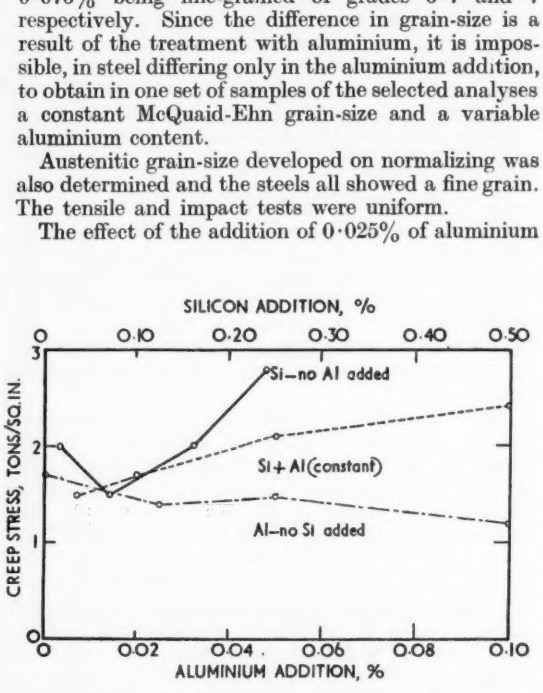


FIG. A—Effect on the Barr-Bardgett creep strength at 450° C. of additions of aluminium and silicon to mild steel

Series 3—Effect of Treating with Aluminium in the Presence of Silicon

Series 3 was composed of four steels, one of which had no ferrosilicon added, the other three having additions of ferrosilicon giving ultimate silicon contents of 0·10, 0·25, and 0·50% respectively; each steel was subsequently treated with 0·050% of aluminium. The analyses, McQuaid-Ehn grain-size, and results of mechanical and creep tests are shown in Table A.

The McQuaid-Ehn grain-sizes of the steels differed only slightly, all being fine-grained. The austenite grain-size of the normalized samples was also fine.

Owing to the influence of silicon on the tensile strength, the steels ranged from 27·6 to 31·9 tons/sq. in. maximum stress. The other mechanical properties were fairly constant.

The steel treated with aluminium and with no silicon added, steel 8, is the same as steel 6 included in series 2. The effect of increasing the silicon content from 0·035 to 0·50% is to increase progressively the creep-stress value from a figure of 1·5 to 2·4 tons/sq. in.

The results on these three series of steels may conveniently be compared by an examination of Fig. A, which shows the creep-stress values plotted against the silicon and aluminium contents, the scales for aluminium and silicon being in an arbitrary ratio of 1 : 5. Comparing the results of series 3 with those of series 1 and 2, it will be seen that the values in general lie intermediate between those of series 1 and 2, indicating that additions of aluminium to the silicon-treated steels have the effect of reducing the creep resistance. The effects of silicon and aluminium are additive. The effect of increasing silicon content, as shown by series 3, is apparently markedly greater than the effect of silicon on the corresponding series in the author's tests. This is again no doubt due to the lower creep resistance of the base composition used in the writer's tests compared with the author's (steel 1 in each case).

In extending this work to basic-open-hearth boiler-plate steels, in order to determine to what extent the effect of silicon, as shown by the experimental steels,

TABLE B—*Analysis and Short-Time Creep-Test Data on Boiler Plates*

Cast No.	Wt. of Si-Mn Added to Ladle, lb.	Analysis (Plates)					Grain-Size (McQuaid-Ehn)	Heat-Treatment Temperature from which Air-Cooled, ° C.	Rate of Creep at 24/48 hr., in./in./hr. × 10^{-6}
		C, %	Mn, %	Si, %	S, %	P, %			
82662	100	0·14$_5$	0·44	<0·010	0·027	0·021	1–3	900	69·3
82758	,,	0·16	0·61	<0·010	0·029	0·038	2–4	900	17·2
82762	,,	0·16	0·63	<0·010	0·041	0·039	1	900	13·7
82690	,,	0·16	0·60	0·010	0·031	0·030	1, 3–4	900	21·8
82689	,,	0·15	0·53	0·010	0·051	0·042	1–3	900	20·8
83006	,,	0·16	0·48	0·015	0·044	0·030	0–1	900	35·0
83285	,,	0·17$_5$	0·51	0·015	0·032	0·041	2–3	900	22·2
82659	,,	0·15$_5$	0·53	0·015	0·034	0·035	2–4	900	20·9
82679	,,	0·20	0·59	0·015	0·029	0·038	2–3	900	12·0
82713	,,	0·17$_5$	0·63	0·015	0·046	0·051	2–3	900	9·0
82817	,,	0·17	0·59	0·020	0·030	0·036	2–3	900	21·5
85726	,,	0·25	0·63	0·025	0·040	0·033	1–3	880	12·5
82646	,,	0·18	0·58	0·030	0·030	0·029		890	22·5
86364	250	0·22	0·69	0·011	0·040	0·046	1–3	890	12·5
86519	,,	0·21	0·60	0·015	0·021	0·037	2–4	900	17·7
86365	,,	0·22	0·67	0·015	0·036	0·045	1–2	890	10·0
86366	,,	0·21	0·62	0·018	0·035	0·045	1–3	890	9·6
86624	,,	0·21	0·62	0·020	0·026	0·037	1–2	900	28·6
85749	,,	0·22	0·63	0·020	0·031	0·040	1–3	880	11·4
86625	300	0·19	0·69	0·030	0·023	0·035	1–2	900	10·8
83471	1200	0·18	0·66	0·132	0·030	0·038		900	6·7

was applicable to commercial steels, it was found that phosphorus has an important influence on creep resistance. The author does not refer to phosphorus as a factor in creep resistance and in his tests this element has probably remained relatively constant, but it is shown to vary in the steels referred to in Table I, which incidentally should read 0·040, 0·028, and 0·017% respectively for steels 75, 35, and 52.

The creep tests carried out on these steels were based on British Standard Specification 1271, 1945, "Proof Test for Creep Quality of Carbon-Steel Plate of Boiler-Plate Quality." This consists of a tensile creep test carried out over a period of 48 hr. at a temperature of 450° C. and with a stress of 8 tons/sq. in., the creep rate being measured between the 24th and 48th hours.

Room-temperature tensile tests and creep tests at 450° C. were carried out on specimens machined from samples cut longitudinally from 1¼ in. thick plates. Analyses, grain-size, and creep-rate figures for the various samples normalized at 880°–900° C., according to carbon content, are given in Table B.

One of the steels had a silicon content of 0·132%, that of the remainder being up to 0·030%. The aluminium addition varied up to 7·8 oz./ton and averaged only 3 oz./ton. The small variation in aluminium addition did not show any definite effect on the creep resistance. In making the comparison between the higher-silicon steel and the lower-silicon steels in the early stages of the tests, quite wide variations in creep resistance were found in the lower-silicon steels, and the tests were extended to further samples. The lower-silicon steels showed on the average a very much lower creep resistance than the higher-silicon steel, which was superior in creep resistance to the best of the low-silicon steels. Further examination of the data indicated a fairly definite relation between the phosphorus content and creep rate as shown in Fig. B, the creep rate increasing with decrease in phosphorous content. The creep rate of the steel containing 0·021% of phosphorus was more than seven times that of the steel containing 0·051%. This difference is, however, partly attributable to variations in the manganese content. The effect of phosphorus in increasing the creep resistance was additive to that of silicon as shown by the points on the graph (Fig. B) for the steels containing 0·025–0·030 and 0·132%, compared with those for the steels with up to 0·020% of silicon.

The author has referred to the applicability of the short-time test as a means of differentiating between steels having widely different creep characteristics in long-time tests. It is of interest to consider the short-period test in relation to the long-time test and in this connection the results of tests carried out in the laboratory with which the writer is associated are presented.

These tests were carried out on steel from cast 83471 containing 0·132% of silicon and on steel from cast 82817 containing 0·020% of silicon. The former is subsequently referred to as the high-silicon steel and the latter as the low-silicon steel. The respective B.S.S. proof tests for creep quality gave creep rates of $6·7 \times 10^{-6}$ and $21·5 \times 10^{-6}$ in./in./hr. The tests were designed with the object of determining the stresses corresponding to a creep rate of 10^{-7} in./in./hr. measured at 1000 hr., but the results permit deductions being made for smaller and larger rates of creep. The results of the creep tests carried out at temperatures of 350°–510° C. on normalized samples (air-cooled from 900° C.) are shown for the high-silicon steel in Fig. C and for the low-

silicon steel in Fig. D in the form of stress/(log rate of creep) curves. With decrease in stress at any temperature the high-silicon steel shows a relatively much larger decrease in creep rate than the low-silicon steel. This results in corresponding curves for the two steels crossing, the points of intersection giving the stresses which produce the same creep rate in the two steels for the various temperatures. These creep rates and temperatures are as follows :

Temperature, °C.	Log Creep Rate, in./in./hr.	Creep Rate, in./in./hr. × 10^{-7}
400	$\bar{7}\cdot6$	3·98
450	$\bar{7}\cdot5$	3·16
480	$\bar{7}\cdot5$	3·16
510	$\bar{7}\cdot0$	1·0

The high-silicon steel shows increasing superiority over the low-silicon steel for rates of creep less than those given above. The curves indicate that for higher rates of creep the low-silicon steel would be superior. It is obvious, therefore, that the short-period test gives the order of merit only for certain creep rates and temperatures.

Comparison between the two steels is shown in Fig. E on the basis of stress and temperature for a creep rate of 1×10^{-7} in./in./hr. The creep strength of the high-silicon steel is superior to that of the low-silicon steel at all temperatures between 350° and 500° C. The difference in creep-stress values decreases from about 1·4 tons/sq. in. at 350° C. to 0·8 tons/sq. in. at 480° C., whilst at 510° C. both steels show the same value of 1·5 tons/sq. in. As previously indicated, if the comparison is based on a higher creep rate, the curves will cross at a lower temperature. However, for stresses corresponding to creep rates of 1×10^{-7} strain per hour and smaller, and for temperatures less than 510° C., the long-time tests show the high-silicon steel to be superior to the low-silicon steel.

On page 503 the author makes an interesting reference to the beneficial effect on creep resistance of low carbon content, particularly in alloy steels, but does not quote any data. In this respect the results given in Table C, obtained in the writer's laboratory on a series of 1·0% chromium, 0·50% molybdenum steels, may be of interest. The steels

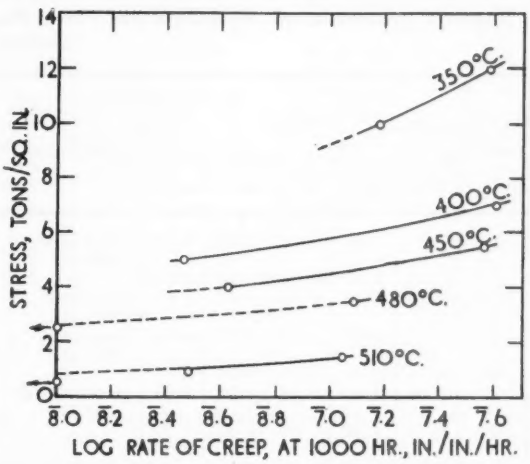


FIG. C—Creep curves for high-silicon plate, cast 83471

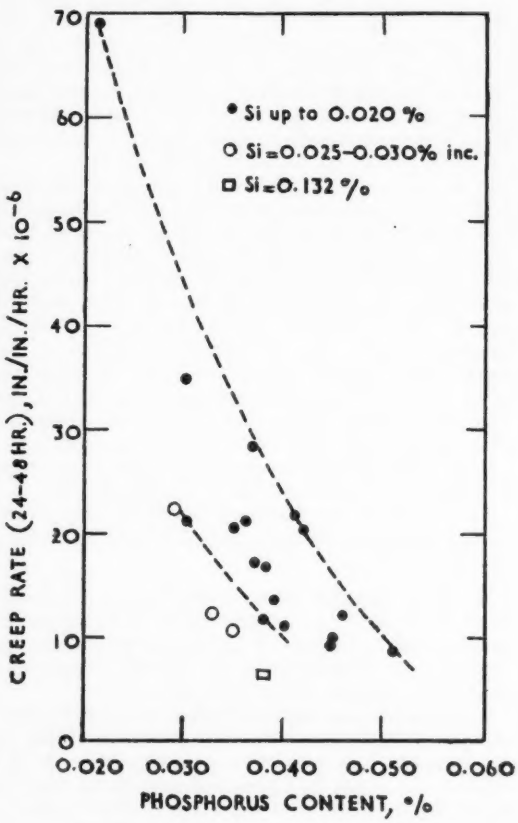


FIG. B—Effect of phosporous and silicon contents on the creep strength of boiler-plate steels at 450° C. (The broken curves show the limits of the range of creep rate for silicon contents up to 0·20%)

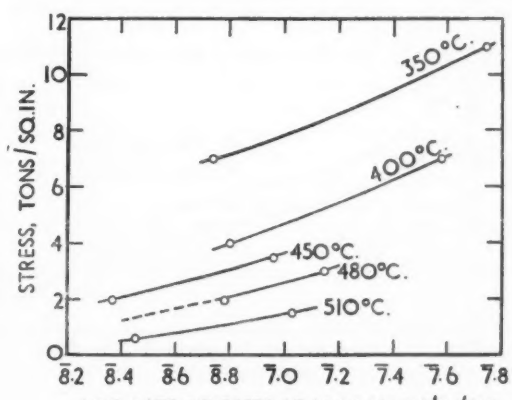


FIG. D—Creep curves for low-silicon plate, cast 82817

TABLE C—*Analysis and Mechanical and Creep Properties of Chromium–Molybdenum Steels of Varying Carbon Content*

Mark	Analysis					Heat-Treatment	Mechanical Properties					Barr-Bardgett Creep-Stress Value at 500° C., tons/sq. in.
	C, %	Mn, %	Si, %	Cr, %	Mo, %		Maximum Stress, tons/sq. in.	Yield Point, tons/sq. %	Elongation on $4\sqrt{A}$, %	Reduction of Area, %	Izod Impact Value, ft. lb.	
22	0·39	0·66	0·18	1·09	0·57	Annealed 865° C.	45·0	27·3	25	42·0	7	2·3
23	0·28	0·62	0·21	1·00	0·54	,,	38·4	25·0	30	49·6	11	2·4
24	0·18	0·60	0·16	1·07	0·60	Annealed 890° C.	32·7	22·6	35	58·2	33	2·3
25	0·12	0·44	0·25	1·13	0·54	Annealed 900° C.	28·7	17·9	39·5	71·6	89	2·9
26	0·08	0·40	0·25	0·96	0·55	,,	26·8	17·9	43	75·6	96	3·0
27	0·08	0·42	0·63	0·78	0·51	Normalized 920° C.	34·4	18·2	37	63·6	47	4·3
28	0·03	0·40	0·56	0·82	0·52	,,	26·7	14·6	43·5	73·6	88	5·0

are not all comparable on account of variation in silicon and manganese contents, but the results show that in the steels containing 0·60% manganese, lowering the carbon content from 0·39 to 0·18% has no appreciable effect on the creep resistance, but further lowering of the carbon content to 0·12% and 0·08% results in a progressive increase in the creep resistance, in spite of the lower manganese content, although the slightly higher silicon content has a beneficial effect. The steel having a silicon content of 0·63% and carbon content of 0·08% shows a marked improvement over the preceding steels of this series. Lowering the carbon content of this steel to 0·03% markedly improves the creep resistance.

Further evidence showing the beneficial effect of low carbon content is given by the creep-stress values in Table D, obtained on a series of 0·50% molybdenum steels. The increase in creep strength with decrease in carbon content occurs in spite of the progressive decrease in manganese and molybdenum contents, which have the reverse effect.

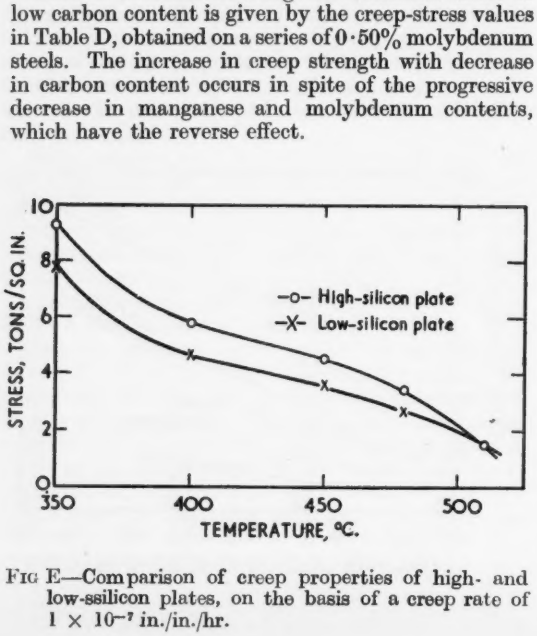


FIG E—Comparison of creep properties of high- and low-ssilicon plates, on the basis of a creep rate of 1×10^{-7} in./in./hr.

On page 511 the author states that aluminium may be used as a deoxidizer provided that the steel remains coarse-grained, but aluminium added to produce a fine grain does not necessarily produce abnormal creep. Further, a coarse-grain steel can have abnormal creep and aluminium additions make the steel more abnormal. It does not follow therefore that one has only to determine whether the grain-size is coarse in order to find out whether a steel has normal creep, and generalizations with respect to grain-size and creep should be considered with reserve.

It would be advantageous in a paper of this type to plot the experimental points and not derive curves such as those shown in Figs. 4–9 from curves such as are given in Fig. 3, as this is liable to lead to unwarranted conclusions and does not give the reader the opportunity of judging the preciseness of the derived curves. For instance, the curve for the 0·40% manganese steels, Fig. 8, if drawn according to the experimental points, would not show a minimum value, but would be of a form similar to that of the remaining curves where the creep rate increases progressively with increase in aluminium additions.

Would the author state how he decides what value of elongation constitutes initial plastic strain ?

The results of the further work which the author is undertaking in this important field will be awaited with keen interest.

Mr. **H. W. Kirkby** (Brown-Firth Research Laboratories, Sheffield) wrote : Mr. Glen is to be congratulated on the excellence of his paper, particularly in regard to the method of attack and the concise way in which the results have been given.

I was glad to see emphasis laid on the actual grain-size *per se* aspect and the care taken in trying to eliminate this as a factor in the investigation. In my opinion, too much has been assumed in the past in regard to grain-size and not enough effort made to distinguish between cause and effect. It does seem, however, that the conditions governing grain-coarsening properties and resistance to creep are interrelated.

TABLE D—*Analysis and Mechanical and Creep Properties of 0·5% Molybdenum Steels of Varying Carbon Content*

Mark	Analysis				Maximum Stress, tons/sq. in.	Mechanical Properties				Barr-Bardgett Creep-Stress Value at 518° C., tons/sq. in.
	C, %	Mn, %	Si, %	Mo, %		Yield Stress, tons/sq. in.	Elongation on $4\sqrt{A}$, %	Reduction of Area, %	Izod Impact Value, ft. lb.	
39	0·05	0·38	0·51	0·51	29·6	16·6	38·5	68·0	86	3·5
40	0·08	0·40	0·56	0·59	32·4	18·7	37·0	66·0	72	3·3
41	0·10	0·47	0·56	0·61	31·9	17·8	27·0	68·0	80	3·1

The influence of silicon and manganese is very interesting, particularly as previous work has more or less ignored these. The earlier work will certainly have to be re-examined.

My main interest in the paper is in regard to the possibilities of a precipitation phenomenon as affecting creep resistance in carbon steels. I have felt for some time that the evidence points in this direction and if I interpret the paper correctly, the same opinion is held by Mr. Glen. To quote, " It is surprising that, apart from Bailey, no one appears to have realized the significance of the thermal hardening, an understanding of which, in the opinion of the author, provides the key to further development of creep-resisting steels." I entirely agree with this statement.

It is very difficult to separate the various factors involved in the creep properties of carbon steel—Mr. Glen has dealt with three, but it is clear, however, that there are more. The effect of aluminium at first suggests oxygen as a factor, particularly if one thinks of a precipitation-hardening phenomenon coupled with, say, strain-age embrittlement and blue brittleness. Aluminium appears to eliminate these latter, or, in other words, to eliminate some type of precipitation hardening. The influence of silicon and manganese, however, complicates matters, as creep resistance decreases with decrease in either or both elements. This would make for an increase in oxygen and any argument on the oxygen theory would have to account for this. It might be that one can have too much oxygen.

The effect of manganese is particularly interesting, especially if one links it with Whiteley's recent grain-size experiments.* I refer particularly to the effect of rate of cooling from " near solidus " temperatures on grain-growth characteristics at 930° C. He showed that by rapidly cooling a coarse-grain type steel, the grain-growth characteristics at 930° C. were considerably modified. In other words, by heat-treatment, a coarse-grain steel could assume at least some of the characteristics of an aluminium-treated fine-grain steel. Slow cooling from such temperatures resulted in the coarse-grained variety of steel.

His further work on ladle and bath samples showed that whereas samples with a manganese content of 0·8–0·9% were sensitive to the rates of cooling mentioned above, when the manganese was 0·05% (and the silicon probably low) the samples behaved as for fine-grain steels, in spite of differences in rate of cooling. Thus manganese would appear to be an important factor and a link-up is suggested with the results obtained by Mr. Glen in regard to the effect of manganese on creep resistance.

It would be interesting to try and correlate the rate of cooling as affecting grain growth in higher manganese material with creep performance, particularly in view of the fact that, according to the literature, the rate of cooling from, say, 1100° C. has a profound effect on the creep resistance of certain aluminium-treated steels. This rate-of-cooling aspect of the matter requires further exploration from all sorts of temperatures and with different compositions. In fact I am wondering what effect would a quicker rate of cooling have on the curves obtained by the author. Is the effect of silicon and manganese to permit a slower rate of cooling for supersaturated solution of say " *X* " ? Does aluminium offset this, so that good creep resistance (and hence correct supersaturation and precipitation) can be obtained only by a comparatively rapid quench ?

I feel that an experiment on low silicon, low manganese, low silicon and manganese, and aluminium-killed variants, with different rates of cooling from 920° C. would help in the general clarification of the problem.

The nature of the precipitating medium is difficult to visualize, but I feel that besides oxygen, we cannot rule out entirely sulphur (as sulphide). A possible indirect link in this direction is the recent American work† on grain-coarsening in terms of aluminium additions and sulphur content. If this is to be believed, then it means that the lower the sulphur content of a steel the more critical is the aluminium addition for producing a maximum coarsening temperature. This is rather against the alumina theory and suggests some relation between aluminium and sulphur. Actually, the formation of an aluminium sulphide critical dispersion has been suggested. I would be interested to hear the author's view in this connection.

* J. H. Whiteley, *Journal of the Iron and Steel Institute*, 1943, No. II, pp. 513P–568P.

† J. W. Halley, *American Institute of Mining and Metallurgical Engineers Technical Publication* 2030 : *Metals Technology*, 1946, vol. 13, June.

AUTHOR'S REPLY

M. J. Glen : I would like to thank Mr. W. B. Brooks for his appreciative remarks and entirely agree with him that further light can be shed on the problem of the effect of aluminium by examination of its effect on ageing. Indeed, the effect of all the elements on creep is related to their effect on ageing.

Mr. Brooks states that suggestive differences will be noticed if various alloying elements are placed in order of increasing effectiveness as solution hardeners and creep retarders, but this does not give a clear picture, since the effectiveness of an element in retarding creep depends on the amount of other alloying elements present.

Mr. W. E. Bardgett's contribution to the discussion is a very welcome and important one. Using the Barr-Bardgett test, he finds that a small addition of silicon decreases the creep resistance, whereas larger additions increase the creep resistance. I did not observe this effect. On the other hand, Mr. Bardgett finds a lesser effect for aluminium. He attributes these variations to a difference in the creep properties of the base composition of the two series of steels. However, further work which I have carried out indicates that such variations are a function of the testing conditions used, *i.e.*, the deoxidizing or alloying effect of an element may increase or decrease the creep resistance depending on the criterion used. Thus, it is only by observing the trend under several sets of test conditions that definite conclusions can be reached regarding the effect on long-time creep properties.

Although I agree with Mr. Bardgett that too high a carbon content is detrimental, it does not follow from Mr. Bardgett's results, showing a 0·03% carbon steel to be the best under his testing conditions, that this carbon content will give the best creep resistance in long-time service, since it is possible by suitable choice of test conditions to show that a comparatively high-carbon steel is the best.

Mr. Bardgett's results on the effect of phosphorus are interesting, but it is significant that the three steels of lowest phosphorus contents also have the lowest manganese contents. If the effect of variable manganese content were eliminated, it is possible that phosphorus would not have such a marked effect.

In the paper abnormality is defined as the deterioration of creep resistance due to the method of deoxidation, and I think that the use of the term abnormality should be restricted to this meaning. Thus, Mr. Bardgett's statement that a coarse-grained steel can show abnormal creep and that an aluminium addition makes it more abnormal simply means that, *e.g.*, a mild steel low in manganese content has a poor creep resistance compared with a steel of higher manganese content. Thus, if a steel is coarse-grained in the McQuaid-Ehn test, it must show a creep resistance which is *normal* for that composition even although the creep resistance is poor.

In the manufacture of the steel used, it was found that manganese was the only element which could not be controlled with reasonable exactitude. In Fig. 3, where manganese is the variable, all the experimental points are plotted and the best curves drawn through these points.

When considering the effect of silicon and aluminium, it would have been misleading to put in the experimental points which varied considerably in manganese content, and thus the derivation of Figs. 4–9 is considered to be the most logical one. In any case, Mr. Bardgett's point is answered in the text, which clearly states that the decrease in creep rate of the 0·4% manganese steel in Fig. 8 may be due to experimental error.

Mr. Bardgett asks what is the exact meaning of initial plastic strain. When a creep test-piece is loaded, the first part of the stress/strain curve is a straight line and if this is extrapolated to the load used the strain value is the elastic extension of the test-piece. The difference between this value of strain and the total strain on loading is the initial plastic strain. This value of course will tend to increase the slower the rate of loading used.

Mr. H. K. Kirkby's extremely constructive remarks are warmly appreciated and will be given full consideration in any further work which is carried out. In particular his reference to the importance of precipitation phenomena and the possible effect of oxygen and sulphur are in agreement with my views, although I believe that it is the solution and precipitation of carbide which is the fundamental fact, since the alteration in creep resistance due to the addition of chromium, molybdenum, or vanadium must be associated with the effect of these elements on the carbides in the steel.

The experiment which Mr. Kirkby suggests on the effect of rate of cooling would be interesting to carry out, but the difficulty is that the result would probably vary with different testing conditions. Thus, I consider that the first requisite for the clarification of creep phenomena is a thorough understanding of the effect of the testing variables, stress, temperature, time, and deformation. When this is known the effects of what might be called the steel variables will then be seen in their true perspective.

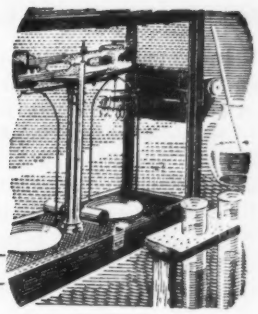
Borax in Steel Wire-Drawing Operations

Dry-Drawing Process

Successful wire-drawing depends largely on the preparatory cleaning of the surface of the metal, and the lubricating film applied to it, prior to die-drawing operations. The usual procedure consists of a preliminary heat-tempering of the metal rods and bars and a dilute mineral acid pickle, followed by an intermediate water wash and, hitherto, by neutralization of the residual acid in a hot, agitated, lime bath. Recent works experience, particularly in the U.S.A., has shown the use of a hot 5–7% borax solution in place of the customary lime (or phosphate) to be a great advance. Easier, cleaner, and safer handling, greater solubility in water (obviating the need for stirring) no necessity for repeated dipping, quicker drying, and lack of flaking, are among the advantages claimed for borax over lime. The suitability of borax as a flux enables the metal stock to be butt-welded for high-speed continuous drawing without the necessity of cleaning the ends, as when lime is used.

B. I. S. R. A.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION



Studies on Anti-Fouling Compositions*

By H. Barnes, B.A., B.Sc., Ph.D., F.R.I.C.†

FOREWORD

The following three papers form part of a series of studies on anti-fouling compositions. The first is a general introduction to the whole series, of which further parts will be published later.

Part II describes the probable reactions occurring between the components during preparation of the varnish, and Part III gives details of a study of the rate of loss of toxic material from a series of varnish media immersed in the sea over two consecutive 40-day periods.

An individual synopsis of each part is given under each title.

Part I—General Introduction

SYNOPSIS

Part I of the paper consists of a general introduction to a study of anti-fouling compositions and their mode of action. Details of the materials and methods used in the investigations are presented. The physicochemical nature of paint is briefly reviewed and the method of approach to the problem is outlined. A summary of the general evidence for the thesis to be developed in subsequent papers is included.

THIS series of papers gives an account of the work carried out by the author on behalf of the Marine Corrosion Sub-Committee. For Part I it was thought desirable to give an introduction and a description of the general experimental methods used.

These studies are largely concerned with an attempt to elucidate the behaviour with respect to one property, namely, the release of contained solid toxic material from a series of anti-fouling compositions to sea-water, since the available evidence suggested this property to be of fundamental importance in the prevention of fouling.‡[1, 2]

In planning the work it was considered that the study of anti-fouling action should, in the first instance, take precedence over other criteria : the primary object of an anti-fouling film is to prevent the settlement of plant and animal forms, and all other properties are of a secondary nature. When the factors governing the anti-fouling action of a composition are clearly understood and defined, the limits can be set within which changes may be made to improve or modify the composition from the purely technological standpoint, and within these limits it will be essential to work.

‡ Simultaneously with this work anti-fouling investigations were being vigorously pursued in the U.S.A., and the American workers published some of their results while the present series of papers was in preparation.

* Paper MG/BC/48/47 of the Marine Corrosion Sub-Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association, received 8th August, 1947. The views expressed in it are the author's and are not necessarily endorsed by the Sub-Committee as a body.

† Formerly Investigator to the Marine Corrosion Sub-Committee of the Iron and Steel Institute, now with the Scottish Marine Biological Association, Millport.

In view of its complex colloidal nature, either as wet paint or as a dry film, it is pertinent to briefly review the nature of the paint system and to consider its relationship to the property under special investigation. Paint is essentially a suspension of pigments in a liquid medium. In its preparation a pigment–air interface is replaced by a pigment–vehicle interface, and the relations at this new boundary are of great importance in determining the physical characters of the product. Further, in the particular systems under investigation there is frequently present an acid resin together with a basic pigment, which react together during the milling process resulting in the production of metal soaps.

When a composition is painted out, the properties conferred on the system by the milling process and the nature of the materials used, have to be considered in relation to factors producing drying, the latter taking place, in the compositions under consideration, largely by the formation of a rigid gel on evaporation of the solvent. The results of the interaction of these factors, within and exterior to the paint, will determine the final state of affairs existing at the ultimate dry film–air interface, which is in turn a determining factor in the initial and early stages of the leaching process. Ultimately, the system to be studied is a dispersion of pigment particles in a solid gel which is itself a complex of rosin, resins, binders, soaps, and probably retained solvent.

In view of this complexity it was thought desirable in the first instance to simplify the system used as far as possible, and to restrict the poisonous pigments to cuprous and mercuric oxides (which general experience has indicated are more efficient than other metal poisons). Attention has been largely concentrated on the former, since for reasons which will appear later, it gives a system which is probably less complex. Red oxide of iron has been used almost exclusively as the inert secondary pigment, for its density, physical properties and particle size are similar to those of the poisonous pigment with which it is employed. (Except when stated as otherwise, it can be assumed that in all of the compositions referred to the secondary pigment is *always* red oxide of iron). As will be shown, it is necessary in a 'normally' pigmented composition* for the vehicle to contain a component susceptible to attack, such as rosin, together with a resistant second component. The simplest type of varnish would therefore be a two-component system containing rosin and inert resin (say, completely esterified reduced rosin) in a solvent. However, in order to obtain an adequate minimum of satisfactory film properties, a third component would of necessity have to be introduced to plasticize such a binary system. In view of this, and in part because much was known of its behaviour from the work of J. E. Harris, the resistant second component of the vehicle has been a Bedesol stand oil (used in the form of a varnish) which, with rosin, gives over a wide range of compositions, a system sufficiently plasticized for the necessary studies. This Bedesol stand oil, which will be termed the binder, has been described as Medium 2 (modified phenol-formaldehyde/stand oil) by F. Fancutt and J. C. Hudson[3] and, unless otherwise stated, it may be assumed that it is the binder component in all the compositions used.

Even with all these limitations and simplifications two rather different approaches are possible. A small number of systems might be investigated with physicochemical precision; this method would involve a careful control of all the conditions during the preparation, the application, and the investigation of the compositions employed. An alternative approach would be to investigate a large number of compositions of widely varying character (within the limits of the simple components mentioned) under comparable but not precisely controlled conditions. In view of the complexity even of the simplest system, the large number of variables—some of which are extremely difficult to control—and in particular the desirability of producing a satisfactory composition, from the practical point of view and within a reasonable time, the latter approach has been used. The study has involved the preparation of approximately 700 compositions, the routine leaching determinations on many of which have been carried on for at least 200 days, thereby involving some 24,000 separate determinations. This method of approach has been justified, since it has shown that successful compositions have rather narrow limits of formulation, and a more intensive study of promising compositions, with a view to their further improvement, and a more precise investigation of the physicochemical factors involved would now be profitable.

In order to save the considerable complications of discussion and cross reference it is desirable to state at this stage the fundamental conception which the interpretation of the results for cuprous oxide compositions is based. It is considered that the major release of copper over much of the composition's life in the types of composition studied, is by direct solution from particulate cuprous oxide. In the early stages this can take place by the direct action of seawater upon such particles as are left on the film surface, but after the removal of such surface

* By a 'normally' pigmented composition is meant one in which the pigment content does not reach the high value required for partial contact throughout the matrix.

particles it involves (in a normally pigmented composition) a breakdown of the matrix by removal of one of the constituent components to allow access to the contained particles. It is believed that diffusion through an unattacked matrix plays little part. This problem of the mode of release of the poison has been the subject of much controversy (*see* the papers by Harris,[2] and Harris and Forbes,[4] and this is not the place to review all the available data ; however, four of the most important lines of evidence in favour of the above hypothesis may be briefly stated. First, all successful compositions contain rosin (or an acidic material) and by partial replacement of this acid rosin by neutral material the rate of loss of copper is gradually reduced and ultimately completely eliminated. (It is true that this could be explained as a gradual reduction in permeability.) Secondly, rosin alone and mixtures of rosin and inert resins containing moderate or high proportions of the former, lose material in sea-water ; the neutral resin content at which loss of material is just prevented in such varnish systems coincides approximately with the point at which loss of copper is prevented when corresponding cuprous oxide varnish systems are examined. Thirdly, a composition which loses copper and prevents fouling does *not* lose copper in a dilute acid citrate solution (in which cuprous oxide is readily soluble). It is unlikely that such a solution would modify the rate of diffusion to any considerable extent, but it is readily explicable on the rosin-solution theory, for the acid nature of the solution would prevent attack of the rosin. Fourthly, painted panels do lose weight on exposure and, as has been frequently confirmed, the loss in weight is greater than can be accounted for by the disappearance of cuprous oxide.

Experimental Methods

Preparation of the Compositions

All the experimental compositions have been prepared in a ball mill. Some of the earlier series were prepared in a simple form of mill in which the jars were rotated by a series of wooden rollers. In later batches a standard Steel-Shaw nine-unit ball mill has been used, 200-g. batches being the usual quantity of each composition prepared. In preparing these paints the rosin was dissolved in the solvent (naphtha) and all the ingredients, *i.e.*, pigments, rosin solution, and binder (also in naphtha), were added at the start ; if necessary the consistency was adjusted by further additions of naphtha in order to obtain correct milling conditions. The standard milling time has been 12 hr.

Certain deviations have been made to investigate special aspects of various problems, but these are indicated at the appropriate points in the text.

Preparation of the Test Panels

As in the earlier work of Harris[2], 7·62 × 2·54-cm. (3 × 1-in.) microscope slides (ground on one side only) have been used and an area of 5·08 × 2·54 cm. (2 × 1-in.) painted. The paint was weighed on application and the rate per unit area was adjusted for the normal type of composition to 0·2–0·25 g. per slide, *i.e.*, 155–194 g./sq. m. In the case of compositions with unusually low or high pigment content the weight was adjusted to give a film of the same approximate thickness. Application in routine work was by a small flat brush, and the slides were allowed to dry indoors for 6 hr. in a horizontal position. No control was exercised over the atmospheric conditions during this period. At least two slides were painted for any one test.

For some of the work a method of application was employed for these small slides, using a doctor-blade technique. Several modifications were investigated but one of the most simple and effective methods involved the use of latex strip. This material (supplied in convenient 1-in. wide strip) was stuck over the surface of the slide, and a window 5·08 × 1·91 cm. (2 × ¾-in.) cut out. Having applied the paint a glass slide resting on the edges of the window was drawn over the surface, and in this way films of definite thicknesses were readily prepared. By varying the number of strips the thickness of the film could be changed by graded amounts.

The Leaching Tests

The simpler modification described by Harris[2] has been used for routine leaching-rate tests, as many as 100 units being run at one time. No temperature control has been employed and fresh sea-water has always been used. In view of the reduced area (one half of that tested in the earlier work) a routine leaching time of 4 hr. has been used, but this was varied for special investigations.

Storage of Slides between Leaching Tests

Except for the investigation of specific theoretical problems, routine test panels were, throughout the work, stored in the sea (*see* Harris[2] ; a discussion of this point will be given in a later paper).

Analytical Methods

The estimation of copper in leachates was always carried out by the extraction technique (Harris[2]), the direct method having been completely abandoned. Mercury in leachates was estimated by the method of H. Barnes,[5] in which interference due to copper is eliminated by the use of potassium cobalticyanide.

Micro methods of paint analysis were developed for use with the small glass test panels and the consequent small paint samples available (H. Barnes[6]).

Leaching in Citric Acid

A sea-water solution of this reagent containing 0·5 g./litre of citric acid has been used, the leaching being carried out in the standard apparatus. In addition, large containers (of 1000 ml. capacity) using 750 ml. of solution have been employed, when the slides were attached to the long inlet tube by means of a rubber band. The agitation of the solution was increased, and since this type of experiment has been used largely for the complete extraction of surface-available copper, as distinct from measuring the rates of solution, this agitation was not very carefully controlled.

Expression of Results

As recommended by the Panel on the Leaching Rate Test[7] of the Marine Corrosion Sub-Committee the leaching rate is expressed in microgrammes per square centimetre per day. Spreading rates throughout are given in grammes per square metre* in conformity with the metric system for the expression of leaching rates ; this method has the advantage that at a given spreading rate and a known leaching rate the possible anti-fouling life can be readily computed. Concentrations are all expressed as milligrammes per litre.

The Materials

The materials were all obtained in fairly large batches so that unless deterioration was pronounced the results over a large number of compositions should be comparable. The Bedesol stand-oil varnish was supplied through the kindness of Mr. Fancutt ; the iron oxide was an I.C.I. product, Indian Red 7 *BS* ; the cuprous oxide (commercial, precipitated for paints) and rosin (Best Amber) were supplied by Messrs. Harrington Bros., Ltd. ; the naphtha was obtained from the Glasgow Corporation Gasworks and termed Water-White high-flash-point (90–190° C.) naphtha.

*1 oz./sq. ft. = 305·2 g./sq. m.

SUMMARY

An account of the methods and materials used in the investigation of a series of anti-fouling compositions is given together with a brief consideration of the nature of the problem and an outline of the thesis to be developed in the subsequent papers of the series.

Part II—The Formation of Copper Soaps in the Preparation of Compositions Containing Cuprous Oxide and Rosin

SYNOPSIS

Compositions used in these investigations contained an acid resin (rosin) and a basic pigment (cuprous oxide), so that in their preparation reaction between these two components takes place, with the formation of copper soaps which are soluble or readily dispersed in the medium. The amount of copper found in the medium is shown to be related to the amounts of rosin and cuprous oxide present and to the milling time used. The possible effect of such changes on leaching and storage characteristics is discussed, and the suggestion is made that the copper content in the medium of paint batches should be determined as part of their routine examination.

In the preparation of compositions using cuprous oxide and a medium containing an acidic component (rosin) in the presence of an organic solvent, reaction between the rosin and the basic oxide would be expected to take place, with the production of a copper soap which is soluble (or readily dispersed) in the solvent used. It has been shown[2] that two of the most important factors regulating the behaviour of an anti-fouling composition are the cuprous oxide content and the proportion of rosin to inert binder in the medium. Since the formation of copper soaps will affect the amounts of both of these components present it is desirable that this process should be investigated, and although this part of the work was interrupted, the results give a clear indication of some of the problems involved and seem worth recording.

The most important influences on the reaction between the rosin and cuprous oxide are brought into play during the milling process, which serves primarily to disperse the pigment. These include temperature changes (local and general), and time and efficiency of grinding (which is itself a whole complex of factors). Under the conditions which are normally used the temperature is not controlled during paint preparation, but for further investigations some form of temperature control is much to be desired. Only the effect of the duration of milling has been investigated using a series of

TABLE I—*Results of Copper Extractions from the Medium of Cuprous Oxide Compositions*

Total pigment content = 67·4%. A dry film basis was used throughout

Composition No.	Cu_2O, %	Rosin in Varnish, %	Copper in 100 gm. of Wet Paint, g.				Equivalent Amount of Rosin Reacted (24 hr. Milling), g.	Rosin Used (24 hr. Milling), %
			Total Milling Time					
			1 hr.	4 hr.	12 hr.	24 hr.		
379	15·6	83·3	0·16	...	0·27	0·37	3·52	16·2
381	31·1	Rosin/binder	0·24	...	0·36	0·48	4·58	21·0
383	46·8	= 5/1	0·33	...	0·45	0·59	5·62	25·8
385	62·4		0·41	...	...	0·70	6·66	30·6
371	15·6	75·0	0·17	0·31	0·39	0·35	3·34	17·2
373	31·1	Rosin/binder	0·28	0·40	0·49	0·48	4·58	23·4
375	46·8	= 3/1	0·39	0·45	0·56	0·57	5·42	27·6
377	62·4		0·50	0·52	0·68	0·64	6·10	31·0
326	15·6	59·9	...	0·13	0·27	0·30	2·86	18·2
384	31·1	Rosin/binder	...	...	0·37	0·44	4·20	26·8
386	46·8	= 1·5/1	...	0·37	0·46	0·50	4·76	30·4
380	62·4		...	...	...	...	...	...
329	15·6	50·0	...	0·16	0·28	0·36	3·42	26·4
289	31·1	Rosin/binder	...	0·26	0·34	0·40	3·82	29·8
291	46·8	= 1/1	...	0·35	...	0·45	4·28	32·6
293	62·4		...	0·40	0·47	0·53	5·04	38·4

compositions milled, as far as could be ascertained, under the same conditions, so that all the other variables were constant. The amount of reaction which has taken place has been estimated by determining the amount of copper in the medium which is soluble in organic solvents.

Immediately after milling, a weighed quantity of the thoroughly mixed paint was taken and the medium separated by repeated additions of chloroform, allowing the precipitate to settle between each addition, and filtering through a No. 42 Whatman filter paper. After making the filtrate up to a known volume, an aliquot part was carefully evaporated, wet ashed, and its copper content determined.

Table I gives the results of copper extracted (grammes per 100 g. of wet paint) for a number of different compositions of variable cuprous oxide, inert filler contents (constant total pigment), and rosin/binder ratio.

Only tentative deductions can be drawn from the limited number of results available, but the following conclusions appear warranted for the compositions examined.

1. With a given medium the amount of copper in the vehicle at any time during the milling process is proportional to the cuprous oxide content of the composition.

2. The milling time has an important effect on the copper content of the varnish. After a short initial period (of rapid reaction) the amount of copper in the varnish is roughly proportional to the milling time (up to 24 hr.) when the rosin content of the varnish is high. With lower rosin contents, the figures suggest that soap formation falls off with time, but the results are somewhat irregular. This is probably due to inadequate control of the numerous variables—particularly the milling speed in relation to consistency—as the components are changed by regular amounts. The results further suggest that in the initial stages of milling there is a very rapid formation of the copper soap or even that some reaction takes place on mixing the components before the milling has been started.

With regard to the effect of this soap formation on the relative proportions, in the finished product, of the materials undergoing reaction, the changes have little effect on the concentration of the cuprous oxide pigment, but the rosin concentration is significantly affected. Whilst there is no doubt that some reaction takes place between the Bedesol stand oil and the cuprous oxide when the two are milled together (a pale green colour is developed), it may be assumed that the major

TABLE II—*Leaching Rates of Compositions 23*H *and 28*H, *Experimental and Commercial Batches*

Composition No.	Copper Leaching Rate, μg./sq. cm./day							
	Soaking Time, days							
	5	15	30	36	60	75	92	98
23*H* (Laboratory)	23·3	29·1	24·7	...	24·4	...	19·4	...
23*H* (Commercial I)	4·9	2·3	...	2·2	...	3·3	...	4·6
23*H* (Commercial II)	2·9	2·8	...	3·8	...	5·8	...	9·9
28*H* (Laboratory)	14·6	25·7	32·8	...	24·2	...	14·9	...
28*H* (Commercial I)	7·5	4·9	...	3·0	...	4·7	...	5·3
28*H* (Commercial II)	5·2	4·8	...	3·1	...	4·6	...	4·1

part of the reaction takes place between the pigment and the rosin. The amount of rosin removed in this way may be calculated on the very tentative assumption that cupric rosinate, $(C_{20}H_{29}O_2)_2Cu$, is formed, and the actual percentage of rosin converted to soap after 24 hr. milling has been calculated on this basis and is shown in Table I.

The reaction has considerably reduced the amount of free rosin present and significantly affected the rosin/binder ratio ; such a change in the percentage of rosin in the varnish can have a profound effect on the leaching rate of a composition. The milling of the compositions should therefore be carefully controlled, since any variation will be reflected in the amount of reaction between the materials used and this will have an important influence on the subsequent leaching rate.

The colour of the medium as a result of the reaction is green, and this is one basis for the assumption that a cupric soap is formed. A cuprous soap can be prepared under rather special reducing conditions but readily oxidizes on exposure to the air. The oxidation which takes place during milling may be due to the direct action of oxygen contained in the mill jar or possibly at the expense of the components of the composition itself. There is also the possibility that at least some of the copper soap arises from reaction between the rosin and cupric oxide or copper present as an impurity in the lower oxide. Messrs. Harrington Bros., Ltd., kindly supplied some details of their product, which showed 0·34% of metallic copper as impurity. Baker and Gibbs[8] give a number of results indicating that various samples of cuprous oxide which they examined contained 0·3–9·7% of cupric oxide and 0·69–9·45% of metallic copper as impurities. Considerations such as these will, however, not affect the argument, since it is pointed out that the change in *rosin* content as a result of reaction is the major consideration.

Variations due to the changed milling conditions in large-scale preparations as compared with those existing in the laboratory experimental work may in part probably account for the different performance of the same composition prepared under the two conditions. Large variations in the leaching rate (and performance) have been observed between batches of the compositions 23*H* and 28*H* formulated by Professor Harris and prepared under the direction of Mr. F. Fancutt in the L.M.S. Railway Co., Ltd., research laboratory at Derby. The leaching-rate data for these are shown in Table II and, whilst such variations may not be entirely due to the changes in milling technique, it is considered that the latter may contribute to the observed differences.

It is suggested that in the preparation of a composition which it is proposed to use on a large scale, the amount of copper in the varnish could profitably be determined both as a routine check and for comparison with the laboratory prototype.

A further point of importance in connection with the reaction during the milling process is its possible effect on the changes taking place during storage. On storing a composition further reaction may take place between the basic pigment and the residual rosin and, since the amount of the latter is dependent on the milling process, the changes on storing may be affected. With efficient milling, any further interaction on storage between pigment and rosin may be expected to be reduced, partly due to the lower rosin concentration and partly due to efficient coating of the particle with soap. The problem of the change of leaching characteristics with storage (by this is meant the changes taking place in a relatively short time, *e.g.*, 100–150 days) was at one time thought to be of considerable importance and some time was devoted to it. The following results can be used to illustrate changes which take place on storage, in compositions which are milled for 5 hr. in the old mill. Such changes can profoundly alter the leaching characteristics and performance : indeed 17*H*, after ageing, has been found on exposure to have lost its anti-fouling properties. However,

TABLE III—*Change of Leaching Rate on Storage (Compositions Milled 5 hr.)*

Composition No.	Cuprous Oxide, %	Age when Painted, days	Leaching Rate, μg./sq. cm./day							
			Soaking Time, days							
			3	7	10	15	34	63	90	120
17*H*	12	3	52	44	...	28	32	17	9	5
21*H*	18	3	30	60	...	31	34	31	21	12
25*H*	24	3	41	70	...	49	49	29	24	15
17*H*	12	120	16	...	18	...	9	8	7	4
21*H*	18	120	15	...	9	...	21	16	12	9
25*H*	24	120	12	...	7	...	19	16	15	8

a comparison of the more recent paints prepared on the Steel-Shaw mill and aged for various periods suggests that much of the change of leaching characteristics was due to inefficient milling on the old mill. The problem of ageing may assume importance when products designed in the laboratory are made on a large scale, for as is well known, equivalent milling efficiency is difficult to attain, and unless the finished paints in both cases are identical the storage properties may not be the same.

This problem of controlled milling becomes more important as further advances are made in formulating an efficient composition, for if the components are regulated so that, when made under given conditions, the leaching rate is *just* adequate to prevent fouling, *i.e.*, for maximum efficiency and anti-fouling life, then should these conditions be changed so that more reaction takes place, the leaching rate of the resultant composition could be lowered and it will no longer be effectively anti-fouling.

In other paint preparations reactive pigments are generally avoided, the sole object of milling being to obtain effective pigment dispersion. Its importance to anti-fouling compositions lies in the fact that the release of poison and the resultant anti-fouling action are controlled primarily by the *chemical* nature of the film, and this can be altered by changes in the milling process.

SUMMARY

Data have been presented to show how copper-soap formation in a series of compositions is affected by cuprous oxide content, rosin content, and milling time. The effect of these changes on the rosin/binder ratio is discussed.

It is suggested that variable milling conditions are in part responsible for the discrepancies in the properties of laboratory and large-scale preparations of the same composition : examples of such discrepancies are quoted. The effect of soap formation on storage changes is discussed. It is suggested that a routine analysis of the copper content of the varnish should be made.

Part III—The Rate of Loss of Material from a Graded Series of Varnishes Exposed in the Sea

SYNOPSIS

The effect of the proportion of rosin to inert binder has been previously shown to have a great influence on the leaching characteristics of cuprous-oxide-containing compositions. A study of the rate of loss of material, in the sea, from varnish media containing graded proportions of rosin and inert binder, has been made. Over the first period (0–40 days) losses were approximately proportional to the percentage of rosin in the varnish. In the second period (40–80 days) the weight losses were much less and seemed to have no clearly defined dependence on the components of the varnish. During this period calcium salts become deposited in the film surface. The inert binder is shown to have a depressing effect upon the rate of loss of material. Evidence is presented to show that the losses observed are due to the removal of rosin, and suggestions are made concerning the mode of action of the rosin. It is shown that acidic material is present in the layers of the film, even after considerable weight losses have taken place.

XUM

Introduction

It has been shown by Harris,[2] and repeatedly confirmed in these investigations, that the constitution of the varnish plays an important part in the release of poison from anti-fouling compositions. It therefore seemed very desirable that some study should be made of the actual rates of solution of rosin and rosin–binder mixtures in order to substantiate and amplify the results on the paints themselves. Information of this kind could be obtained by a leaching technique if a sufficiently sensitive quantitative reaction for rosin could be devised. Such a technique would have the additional advantage that the rate of loss could be measured both in pigmented and in unpigmented material, and that any variation of the rate of loss with duration of exposure would be clearly indicated. In the absence of such a method it was felt that some pertinent information might be obtained by measuring the weight losses of panels coated with varnishes and exposed in the sea. Perhaps the most unsatisfactory feature of this procedure is that it only gives the mean rate of loss over quite an extended period, for the exposure period must be sufficiently long to give measurable weight changes.

Table IV—*Loss in Weight of Rosin–Hydrogum Series for 40 days Exposure*

Rosin, %	Loss in Weight, mg./12·9 sq. cm.	Mean Rate of Loss, μg./sq. cm./day
87·5	52·8 48·7 52·6	100
83·3	40·6 40·5 44·1	81
80·0	37·4 37·1 35·7	71
75·0	29·1 30·5 31·8	59
71·4	21·8 22·7 19·5	41
66·6	13·1 13·4 14·0	26
50·0	0·2 3·7 0·8	3
33·3	3·8 2·7 —	6

Experimental Methods and Materials

Preliminary experiments had indicated that with rosin and rosin–binder mixtures containing large proportions of rosin, the rate of solution was high and that reasonably thick films would be necessary. In order to use the form of carrier normally employed for the exposure of leaching-test panels 7·62 × 2·54-cm. (3 × 1-in.) microscope slides were used. The rate of loss of material may change with time but the exposure had to be of sufficient duration to give significant weight losses which could be measured on an ordinary analytical balance. The use of thick films involves the application of several coats, and preliminary experiments indicated that a very considerable amount of solvent was retained under these conditions. (Difficulties due to retained solvent had also been experienced in experiments using buffer solutions to investigate the rate of solution of rosin-containing varnishes.) In view of the possible effect of this retained solvent on the dissolution of the varnish, it was decided to bake the films to ensure its removal. The danger of such a process was fully appreciated—the possibility of oxidation and other changes—but it does remove the retained-solvent difficulty, and the series throughout is comparable. The actual rates observed are of the same magnitude as indicated by paint studies.*

The varnish films are non-toxic and would, therefore, become covered by various forms of fouling if exposed during the fouling season. The period of exposure of the films for experimental work of this kind is therefore limited to the winter months of December to March, when few fouling organisms are present and during which time the sea temperatures are 5°–6° C. in contrast to the higher summer temperatures of approximately 12° C.

Two-component (excluding solvent) varnish systems were used throughout, no additional plasticizer being added. Varnishes containing rosin alone and rosin in combination with eight gradually increasing additions of binder were used, giving 87·5, 83·3, 80·0, 75·0, 71·4, 66·6,

* In view of the experimental technique adopted it is realized that great caution must be exercised in applying the results to a paint which, in addition to containing other components which might modify its behaviour, is applied as a thinner film and allowed to dry in air.

50·0, and 33·3% of rosin, corresponding to rosin/binder ratios of 7/1, 5/1, 4/1, 3/1, 2·5/1, 2/1, 1/1, and 0·5/1, respectively. Four different binder components were used; the usual Bedesol stand oil (of acid value 15), ester gum (glyceryl ester of rosin, acid value 11·4), hydrogum (hydrogenated esterified rosin, acid value 5·5), and Beckacite 100 (an esterified reduced phenolic resin, acid value 17). All of the varnishes were prepared by dissolving the components in naphtha, mixing in the required proportions, and adjusting the viscosities to a roughly comparable value. (In all of the varnishes the total solid content was approximately 70% by weight.)

The varnishes were applied, with a flat 2·54-cm. (1-in.) wide brush, to 12·9 sq. cm. (2 sq. in.) of a microscope slide ground on one side only. Several coats were applied and were allowed to dry between each application, until a thickness of 0·4–0·5 g. of varnish was obtained on the slide. After the final coat had dried overnight, the edges were trimmed and thoroughly cleaned by means of a volatile solvent (chloroform), and the slides were then baked at 150° C. in an electric oven for 3 hr. in a current of nitrogen. The slides were allowed to cool for a short time in the oven and then in a dessicator. They were then weighed on an analytical balance.

After weighing, the backs and the unpainted portions were wrapped with latex strip to keep them free from rust (a precaution dictated by experience), and fitted in the usual leaching slide carriers. As two exposure periods were planned and each varnish was exposed in triplicate a total of 192 slides were exposed in this way. Whilst on the raft the slides were periodically inspected; during the first exposure period of forty days no fouling organisms settled, but a number of slides were lost.* Towards the middle of the second exposure period there was a considerable amount of diatom settlement; the carriers were therefore brought into the laboratory and, without the removal of the slides, were kept for 12 hr. in sea-water containing a small quantity of formalin brought to *p*H 8·0 by sodium borate. The diatoms were then removed by very gentle brushing with a small paint brush. On replacing the slides on the raft they were immersed at a lower depth, and no further trouble was experienced.

On completion of the exposure periods the slides were removed from the raft, carefully cleaned from latex, and washed in running tap-water for three days, the surfaces being *very lightly* brushed to remove traces of rust and debris. After drying in air for a day and in a desiccator for three days they were weighed, the loss in weight being used to calculate the mean rates of loss over the exposure period.

The Results

It is hardly to be doubted that the rate of loss changes during the exposure. A preliminary survey indicated that the losses during the second exposure period were quite different from the first,

* Owing to rough weather this varnish series suffered more slide losses than usually experienced, thus accounting for the incomplete results found in some of the following Tables.

TABLE V—*Loss in Weight of Varnishes for 40 days Exposure*

Binder	Rosin, %	Mean Loss in Weight, mg./12·9 sq. cm.	Rate of Loss, μg./sq. cm./day	Binder	Rosin, %	Mean Loss in Weight, mg./12·9 sq. cm.	Rate of Loss μg./sq. cm./day
Nil	100·0	88	171				
Bedesol Stand Oil	87·5	59	114	Beckacite 100	87·5	46	89
	83·3	47	91		83·3	36	69
	80·0	...	...		80·0	32	62
	75·0	21	41		75·0	31	58
	71·4	12	24		71·4	27	51
	66·6	3	6		66·6	21	41
	50·0	0·1	0·2		50·0	4	8
	33·3	+2	3		33·3	0·5	1
Ester Gum	87·5	63	122	Hydrogum	87·5	51	100
	83·3	...	...		83·3	42	81
	80·0	56	109		80·0	37	71
	75·0	47	91		75·0	31	59
	71·4	39	75		71·4	21	41
	66·6	32	61		66·6	14	26
	50·0	11	21		50·0	1	3
	33·3	0·6	1		33·3	3	6

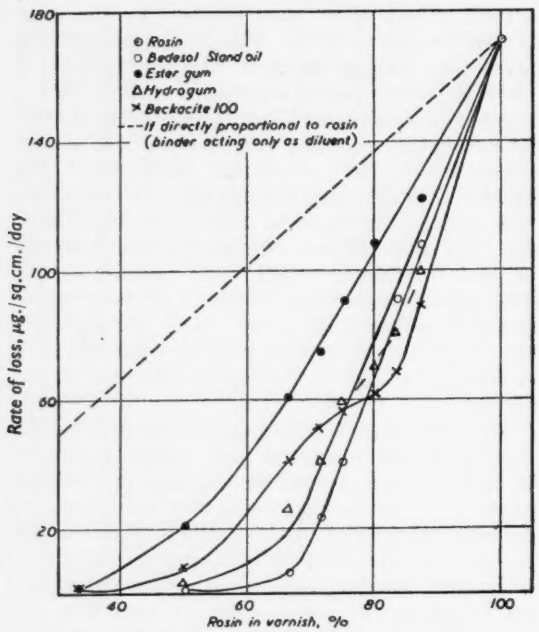


FIG. 1—Rate of loss of material from varnishes

and the results for the two exposures are considered separately. Such a division, which is arbitrary if the rate is changing continuously cannot, however, be avoided (*see* p. 594).

First Exposure Period (0–40 days)

In view of the treatment of the slides the agreement between replicates was remarkably satisfactory : the magnitude of the variations may be illustrated by reference to the rosin–hydrogum series, the results of which are shown in Table IV. The small weight losses at lower rosin content are, in view of the experimental methods, of doubtful significance.

The results for the whole series of varnishes are shown in Table V and Fig. 1.

It is clear that the addition of the binder depresses the rate of solution very considerably. With the possible exception of Beckacite 100 the rate of solution is proportional to the rosin content down to about 65% of rosin ; below this value the losses are very small and the relation is no longer linear ; at about 30% of rosin content the loss is completely eliminated.

The relation between rosin content and rate of loss of weight, together with the fact that at 70% of binder content no loss takes place, suggests that the material undergoing solution* is the rosin or some component of it. The rosin is distinguished from the binder component by the fact that it is largely composed of acidic material, although the nature of the component acids and quantitative proportions in the commercial product are only incompletely understood. The evidence available indicates the presence of a high content of acids isomeric with abietic acid and all may be represented by the formula $C_{20}H_{20}O_2$. In view of the alkaline nature of the sea-water, it is suggested that it is these acid components which are dissolved.* It should be remembered, however, that rosin contains in addition inert non-saponifiable material (14% in the present instance, as indicated by the acid value) which is usually considered to be of a diterpene nature.

Strict proof that it is only these acidic components of the rosin in varnishes which dissolve would require the isolation and identification of the material lost and an agreement between the weight losses observed and the amount recovered. Such proof is clearly not possible under the experimental conditions used, but some evidence on this point is presented later (*see* p. 597).

The acidic nature of the rosin component will explain qualitatively why the weight losses should be obtained. It is clear, however, from the figures, that if the above assumptions are correct, the inert rosin has a depressing effect on the observed rate of solution ; if the binder were only acting as a diluent then the rate of loss would be directly proportional to the rosin concentration (*see* Fig. 1). The complex nature of the solid phase and the fact that even with pure materials the kinetics of heterogeneous reactions are those least understood, make any suggestion as to the mode of action of the inert binder of a very tentative nature. It seems probable that in such organo-gels which these varnishes must represent (whether the materials are truly compatible or not) considerable intermolecular forces are involved, leading in many cases to orientation of the constituent molecular species. Any such forces or orientation could tend to reduce the rate of solution of the contained acids. It is conceivable, therefore, that the activity of the acids and, hence, the rate of solution of the molecules will vary according to their position in the structure of the gel.

If any complete 'inactivation' takes place then the skeleton left on the surface of the film should contain a measurable amount of rosin acids, and an attempt has been made to obtain some information on this point in the following way. The surface of the film was scraped as evenly as

* It is possible that in addition to chemical attack the varnishes may be subject to bacterial action. Such attack might be expected to be least during the winter and in any event the rosin would again be the most readily utilized component.

* This does not preclude other changes—such as oxidation—from taking place, but such oxidative changes would not affect the theory that saponification is largely responsible for solution.

TABLE VI—*Estimation of Composition of Successive Layers of Residual Skeleton*

Original Composition : Rosin Acids, 75·3% ; Resenes, 12·3% ; Binder (Bedesol), 12·4%

Order of Samples from Surface Downwards	Wt. of Sample per Slide (12·9 sq. cm.), mg.	Equivalent in 0·0077*N* of NaOH per 100 mg., ml.	Calculated Composition of Sample			Acid Lost, (per 12·9 sq. cm.), mg.
			Rosin Acids, %	Resenes, %	Binder, %	
1	4·60	14·4	30·5	34·5	35·0	8·3
2	4·35	11·3	23·0	38·0	39·0	7·1
3	5·55	10·5	21·0	39·0	40·0	12·1
4	4·45	13·0	27·0	36·0	37·0	8·6
5	4·20	13·2	28·0	36·0	36·0	8·1
6	3·50	15·2	32·5	33·5	34·0	6·1
7	3·45	17·8	39·0	30·0	31·0	5·0
8	2·50	25·6	58·0	21·0	21·0	1·8
9	2·05	28·5	65·0	17·0	18·0	0·9
10	2·55	28·8	66·0	17·0	17·0	1·0
11	3·05	29·2	66·5	16·5	17·0	1·1
12	3·15	32·7	75·0	12·0	13·0	0·0

possible with a sharp knife, taking off a few milligrammes (corresponding samples from two slides being mixed for titration). This was repeated several times, each layer being weighed, dissolved in a 50% alcohol–benzene mixture, and titrated with approximately *N*/100 sodium hydroxide, using a micro-burette and phenolphthalein as indicator. In calculations relating to the composition of these layers the amount of inert material in the rosin must be taken into account, since although small (14%) it will not dissolve, and forms, therefore, an appreciable proportion of the residual film ; from the acid value of the rosin* the amount of this inert material (which for brevity will be termed 'resenes' and assumed to have a negligible acid value) can be calculated. The composition of the material scraped off can be calculated in the following manner :

Let A = Amount of rosin acids present in 100 mg. of scrape.

R = Amount of resenes present in 100 mg. of scrape.

B = Amount of binder present in 100 mg. of scrape.

Then in 100 mg. of scrape :

$$A + R + B = 100.$$

Also :

$$\frac{A}{100} \times (\text{acid value})_A + \frac{B}{100} \times (\text{acid value})_B = \text{acid value of the scrape.}$$

Since the proportion of resenes to binder is assumed to undergo no change from the original, R can be expressed in terms of B in the first equation and the composition of the scrape determined. Further, the acid originally present in any particular layer analysed in this way can be calculated from its binder content and the original rosin/binder ratio, and hence the amount of acid lost can be computed.

The method of sampling the film is imperfect but it is probably satisfactory as a first approximation. Table VI shows in detail the results of an experiment of this type for one rosin–Bedesol combination.

The figures in Table VI indicate, first, that the outer zone of the surface has a considerable proportion of titratable acid remaining, thus supporting the suggestion made previously that some of the acidic rosin is bound up by the binder and not capable of solution. This amount is fairly constant for the first few layers scraped off and constitutes about 20% of the residual film. Secondly, the total amount lost per slide calculated from these results, namely, 60 mg., agrees very closely with that found by weighing the slides before and after exposure. The assumption that the loss in weight is due to solution of acid components is verified. The agreement between the calculated and observed losses in this instance is very good, undoubtedly due to a cancellation of

TABLE VII—*Calculated and Observed Weight Losses in Rosin–Bedesol Varnishes*

Rosin in Varnish, %	Mean Loss in Weight, mg./12·9 sq. cm.	
	Observed	Calculated from Scrapes
87·5	60	59
83·3	57	58
75·0	21	25
71·4	12	10
66·6	3	0·3

* In what follows, the acids of the rosin are considered as a single entity characterized by the actual acid value : it is realized that the problem is really more complex since it is possible that the different acids may be dissolved at somewhat different rates.

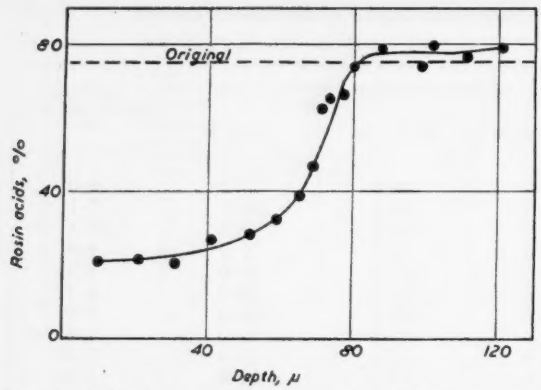


FIG. 2—Calculated percentage of rosin acids present at various depths of film after 40 days soaking

errors. However, analyses of the material scraped from other test slides exposed gave similar results, and on the whole the hypothesis is reasonably substantiated. All the results obtained from the titration experiments of the rosin–Bedesol varnishes* are given in Table VII.

A further calculation of interest can be made from these results. By using the figure for the residual binder content the depth to which the sample had been taken can be calculated, and in Fig. 2 this is plotted against the acid content of the scrape for one rosin–binder combination. If the scrapes approximate to uniform samples over the whole area of the slide and if the attack is uniform over the whole of the film, then the results indicate that in 40 days, up to a depth of 30μ, all the removable acid has been dissolved; under this 30μ is a zone some 50μ thick where solution is taking place. This supports the suggestion that a portion of the acid is more readily dissolved and disappears first, to be followed by the more slowly reacting portion.

* Of all the exposures made the rosin–Bedesol varnishes were most suitable for this type of experiment since their surfaces were not wrinkled and regular scrapes could be taken.

TABLE VIII—*Original and Final Composition of Outer Layers of Varnish Films*

Original Composition			Final Composition		
Binder, %	Resenes, %	Rosin Acid, %	Binder, %	Resenes, %	Rosin Acid, %
12·50	12·25	75·25	39	38	23
16·70	11·66	71·64	46	32	21
25·00	10·50	64·50	53	22	25
28·60	10·00	61·40	51	18	31
33·40	9·32	57·28	35	10	55

The relation between the binder content and the acid remaining in the outer layers of the film is shown in Table VIII.

This indicates that at low binder contents the proportion of inert material to acid in the film surface is constant, with rather more than 20% of acid present.

Second Exposure Period (40–80 days)

The loss in weight during 80 days, taken in conjunction with the first series of exposures, enables the loss in weight to be calculated for the

TABLE IX—*Loss in Weight of Varnishes for the Second Series of 40 days Exposure*

Binder	Rosin, %	Mean Loss in Weight, mg./12·9 sq. cm.	Rate of Loss, μg./sq. cm./day	Binder	Rosin, %	Mean Loss in Weight, mg./12·9 sq. cm.	Rate of Loss, μg./sq. cm./day
Nil	100·0	47	90				
Bedesol Stand Oil	87·5	30	59	Beckacite 100	87·5	17	32
	83·3	28	54		83·3	5	9
	80·0	...	...		80·0	6	12
	75·0	21	41		75·0	5	10
	71·4	6	11		71·4	4	9
	66·6	10	19		66·6	6	12
	50·0	+3	+6		50·0	4	8
	33·3	+2	+4		33·3	+5	+10
Ester Gum	87·5	14	27	Hydrogum	87·5	12	22
	83·3	...	...		83·3	21	40
	80·0	20	38		80·0	18	34
	75·0	17	33		75·0	11	20
	71·4	10	19		71·4	4	7
	66·6	8	15		66·6	7	14
	50·0	7	14		50·0	+1	+2
	33·3	+2	+5		33·3	+1	+1

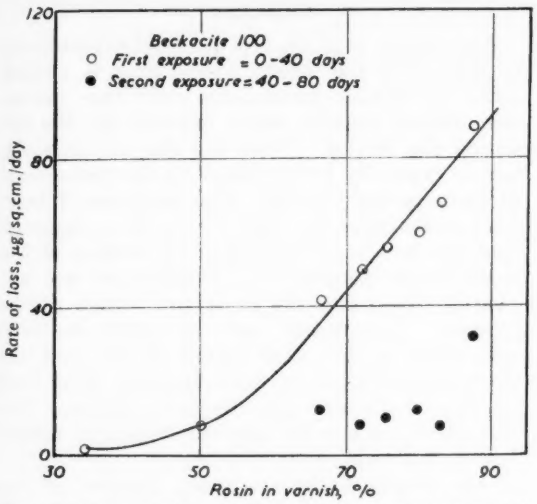


FIG. 3—Rate of loss of material from Beckacite 100 varnish

second period. The results are shown in Table IX and are plotted with the corresponding values for the first exposure period in Figs. 3–6.

It is clear that there is a very marked reduction in the rate of loss during the second 40 days. It was thought that this might be in part due to the formation of a gelatinous layer of insoluble calcium and magnesium salts of the rosin acids in the inner regions of the film, for solution during the second period is, in many of the films, taking place from a very considerable depth.

Qualitative tests had indicated the presence of calcium in the films, and in several of the series the surface was scraped down to the unchanged varnish, the material was ashed, the ash dissolved

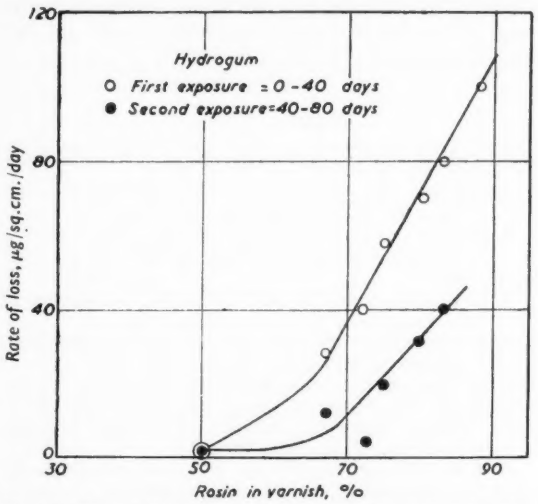


FIG. 4—Rate of loss of material from hydrogum varnish

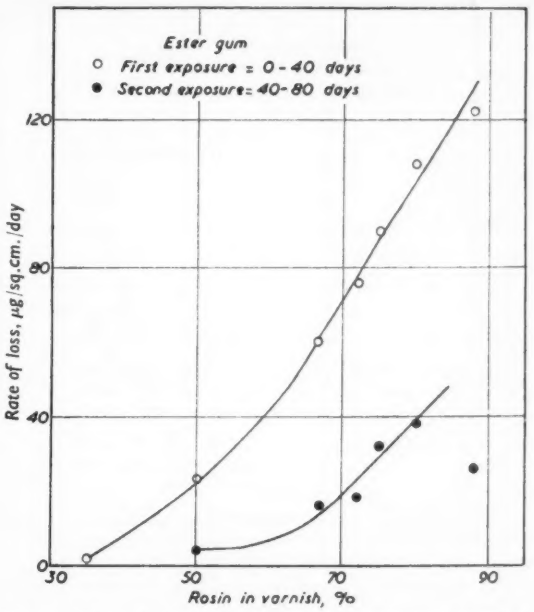


FIG. 5—Rate of loss of material from ester-gum varnish

in dilute acid, and the calcium in the ash deter. mined (after removal of traces of iron originating presumably from traces of adherent rust particles) by a single precipitation as oxalate and subsequent titration with potassium permanganate. The results are shown in Table X.

TABLE X—*Calcium Content of Varnish Films (80 days Exposure)*

Binder	Rosin/ Binder Ratio	Rosin, %	Calcium, mg./12·9 sq. cm.	Equivalent Rosin, mg.
Nil	Pure rosin	100·0	0·50	7·6
Ester Gum	4/1	80·0	0·50	7·6
	3/1	75·0	0·45	6·8
	2·5/1	71·4	0·25	3·8
	2/1	66·6	0·20	3·0
	1/1	50·0	0·10	1·5
	0·5/1	33·3	0·00	0·0
Hydrogum	5/1	83·3	0·15	2·3
	4/1	80·0	0·10	1·5
	3/1	75·0	0·10	1·5
	2·5/1	71·4	0·05	0·8
Beckacite 100	7/1	87·5	0·30	4·6
	3/1	75·0	0·20	3·0
	2·5/1	71·4	0·20	3·0
	2/1	66·6	0·25	3·8
	1/1	50·0	0·15	2·3

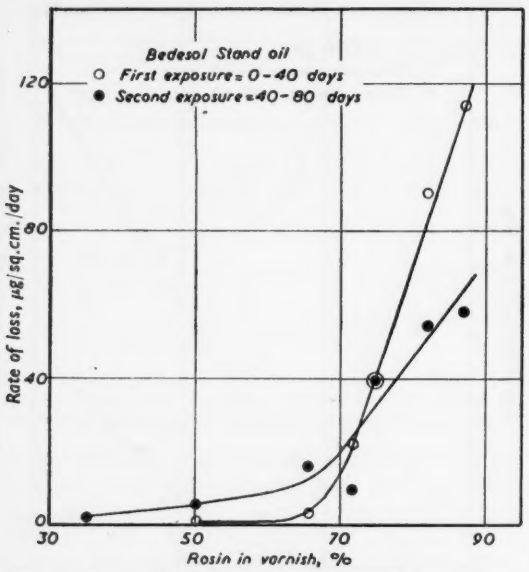


FIG. 6—Rate of loss of material from Bedesol stand-oil varnish

There is some deposition of calcium in the film, doubtless as calcium salts of the rosin acids. In the rosin–ester-gum series the amount of rosin in the film is correlated with the amount of calcium found, but in the others the most that can be said is that there is a tendency for high-rosin varnishes to contain larger amounts of calcium in the film. On the basis of $Ca(Ab)_2$, the amount of calcium rosinate formed can be calculated, and the amount in most cases is sufficient to form a layer of gelatinous material which would hinder solution from the lower layers. Further, there is a tendency during the later stages of exposure for the rate of loss of material from all of these films to approach a low value which is independent of the amount of rosin in the film, which suggests that a layer of liquid saturated with rosin in solution is being formed deep down in the film and that the limiting factor is, then, the rate of diffusion through this impeding layer. It is possible, however, that the strain in the films in the early stages contributes to a higher rate of loss.

The differences between the two exposure periods may reflect a moderately sharp change in the process controlling the rate of loss. Whether this change is immediate or takes place gradually throughout the first 40 days can only be decided by further experimental work which should aim at a determination of the rates of loss over much shorter intervals of time. A leaching technique for this purpose is much to be desired. The part played by bacteria in the weight losses should also be elucidated.

SUMMARY

An account is given of a series of experiments to determine the weight losses from a graded series of varnish components over two successive 40-day periods, when exposed in the sea during the winter. Over the first 40 days the loss is generally proportional to the percentage of rosin in the varnish. The presence of inert binders depresses the rate of loss. It is suggested that the loss is probably due to solution of the acidic rosin components. Suggestions are advanced concerning the protective action of the binder. Experiments on the acidic material remaining in the outer layers of the film are described, and calculations indicate that acid material is 'bound' in the outer layers. The calculated loss is in fair agreement with the weight losses observed.

The weight losses are much reduced in the second exposure period (40–80 days). During this period it is shown that calcium salts are deposited in the film matrix.

The desirability of devising a leaching technique as well as for the study of the rôle of bacteria in the solution of these films is stressed.

ACKNOWLEDGMENT

The author wishes gratefully to acknowledge the interest shown by and the valuable criticisms of Professor J. E. Harris during the whole course of this work.

The compositions referred to in Part II were made by Mr. Wilkie of the Admiralty Chemical Department, Portsmouth. Mr. Wilkie was temporarily transferred to work with the author while the investigations now recorded were in progress. The author wishes to record his gratitude to Mr. Wilkie for his assistance, particularly in the preparation of compositions and in the leaching-rate work.

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The Design of Open-Hearth Gas Ports*

By M. P. Newby, B.Sc., A.Inst.P.†

SYNOPSIS

Experiments were carried out on a model of the gas port of a producer-gas-fired open-hearth furnace with the object of determining the efficiency of a port considered as an arrangement for converting pressure energy into kinetic energy of gas motion; the resistance of the port to the outgoing products of combustion was also measured. The experiments showed that the designs tested have fairly high efficiency and that changes in design can bring only a limited improvement. Funnels added to the port between throat and furnace chamber gave considerable reduction of outgoing resistance but involve practical difficulties which make them of limited industrial value, and it is concluded that the only way to increase the gas velocity substantially is to increase the pressure behind the port.

INTRODUCTION

THE designs of air and gas ports in the open-hearth steel furnace are of fundamental importance in that they directly influence the flame characteristics. Combustion in the flame depends upon the intimate mixing of air and gas and, as Rummel[1] has shown, this depends upon the eddy diffusion between the two. He has also shown that increased rate of mixing can be attained by increasing the velocity of both air and gas streams and also by increasing the difference in velocity between the two.

The furnace trials recently reported by Chesters and Thring[2] and the model-furnace trials reported by Leckie and his co-workers[3] confirm that the mixing is influenced by the shape of the air and gas ports and by the position of the ports in relation to the furnace and to each other. These trials, in conjunction with the theoretical examination of the heat transfer in the open-hearth furnace,[4] show that rapidity of mixing is advantageous in that it ensures a high flame temperature early in the passage of the gases through the furnace chamber. This temperature should be as high as possible without causing damage to the refractories, and in particular the roof refractories.

Aim of the Experiments

The ultimate aim of the present series of experiments is to investigate by the use of aerodynamic models all the factors influencing the mixing of the gas and air streams within the furnace. The experiments here discussed have, however, been carried out with the limited objective of determining by what means the gas velocity on entry to the furnace chamber may be increased. For this purpose measurements have been made of the pressure drops occurring in gas ports.

The pressure losses to be expected at bends and throats have been previously determined,[5,6] but it was not known how the overall losses were affected by the close combination of bend and throat which constitutes a gas port.

THEORETICAL CONSIDERATIONS

(a) *Energy Changes in Port*

The gas port is a device for introducing gas into the furnace and imparting velocity to it at the same time. If q is the quantity of gas required in cubic feet per second, then $q = V_U A_U = V_T A_T$, where V_U, V_T, are the velocities of the gas in uptake and port throat respectively, and A_U, A_T are the cross-sectional areas at the same points. A smaller throat area thus corresponds to a higher velocity of entry to the furnace for a given quantity of gas per second.

Let p, ρ, and V be respectively the pressure, density, and velocity of the gas and h the vertical height, and let the suffixes T and U stand for the throat and a fixed point in the uptake, respectively. If Δp be the total pressure drop between uptake and furnace, including buoyancy pressure, then:

$$\tfrac{1}{2}\rho(V_T^2 - V_U^2) = \Delta p, \quad \text{.....(1)}$$

or if pressure is expressed in terms of pounds per square foot instead of poundals per square foot:

$$\rho(V_T^2 - V_U^2)/2g = \Delta p \text{ lb./sq. ft.,} \quad \text{.....(2)}$$

$$= \frac{\Delta p}{5{\cdot}2} \text{ in. W.G.,} \quad \text{.....(3)}$$

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* Paper SM/A/56/47 of the Steel Practice Committee of the Steelmaking Division of the British Iron and Steel Research Association, received 24th July, 1947. The views expressed are the author's and are not necessarily endorsed by the Committee as a body.

† British Iron and Steel Research Association.

TABLE I—*Dynamic Heads of Air*

Temperature, °C	0	100	200	300	400	500	600	800	1000	1500	2000
Dynamic head, in. W.G. × 10^{-4}	2·41	1·76	1·39	1·15	0·977	0·851	0·753	0·613	0·517	0·371	0·289

where $\rho(V_T^2 - V_U^2)/2g$ is the kinetic energy per unit volume imparted to the gas by the port.

The dynamic heads, $\rho V^2/2g$, in ten-thousandths of an inch water gauge for air moving at one foot per second at normal atmospheric pressure are given in Table I. For any other gas the figures given must be multiplied by ρ-gas/ρ-air, the density of the gas relative to air, and for any other velocity they must be multiplied by V^2, where V is measured in feet per second.

(b) *Frictional Effects*

If frictional effects exist in the flow system the kinetic energy developed will be less than that given by equation (3). The pressure drop in such a port will consist of two parts, one of which is used to produce kinetic energy and the other to overcome friction. This frictional pressure drop will occur largely through the formation of eddies, which are eventually destroyed by the viscosity of the gas.

If the pressure drop used in producing kinetic energy be termed Δp_e and the pressure drop used in overcoming friction Δp_f, we have :

$$\rho(V_T^2 - V_U^2) = \Delta P - \Delta p_f = \Delta p_e, \quad \text{.....(4)}$$

where ΔP is the total pressure drop. This equation may be written :

$$2g\Delta P/\rho(V_T^2 - V_U^2) = K, \quad \text{.....(5)}$$

where K is greater than unity.

(c) *Means of Increasing Kinetic Energy of Gas*

The relation between ΔP and $\rho(V_T^2 - V_U^2)/2g$ will be considered from another aspect.

Let us consider the flow of fluids in two geometrically similar ports. The two flows are dynamically and geometrically similar, provided that the Reynolds number of the flow is the same at corresponding points of the two systems. The Reynolds number, R, is equal to VL/ν, where ν is the kinematic viscosity of the gas and L a characteristic length of the system ; R is a dimensionless quantity, and any other dimensionless quantity characteristic of the flow will be a function of R. The expression $2g\Delta P/\rho(V_T^2 - V_U^2)$ is such a dimensionless quantity, and we may therefore write :

$$2g\Delta P/\rho(V_T^2 - V_U^2) = \phi(R). \quad \text{.....(6)}$$

This formula will be applicable equally to a full-size furnace port and to an exact scale model, and holds for liquids and gases of all viscosities.

We may anticipate the experimental results by stating that $\phi(R)$ is found to be constant over a great range of R for such high values of R as occur in furnace practice. Comparing equations (4) and (5) we see that :

$$2g\Delta P/\rho(V_T^2 - V_U^2) = K = \phi(R), \text{ a constant} \quad \text{.....(7)}$$

As $\rho(V_T^2 - V_U^2)/2g = \Delta P/K = \Delta P - \Delta p_f$, we may increase the kinetic energy in the port throat either by increasing the total pressure drop ΔP or by decreasing K. The value of K, however, equals $\Delta P/(\Delta P - \Delta p_f)$ and may only be reduced by reducing $\Delta p_f/\Delta P$. Only a limited increase in kinetic energy is thus possible by reducing the frictional pressure loss Δp_f, and any further increase must be obtained by increasing the total pressure drop across the port.

The experiments carried out were directed to finding the value of K for various designs of port and in ascertaining whether there are practicable means of bringing that value nearer to unity.

(d) *Friction Factor*

In order to facilitate the comparison of different ports the quantity K has been slightly simplified to $F = 2g\Delta P/\rho V_T^2$. The expression F is generally called the friction factor, and it is based on the highest value of the gas velocity in the section of flow under consideration ; F and K are connected by the equation :

$$F = [1 - (A_T/A_U)^2]K \quad \text{.....(8)}$$

(e) *Flow Through Orifices and Nozzles*

In the above discussion the cross-sectional area of the throat has been taken as the smallest in the flow system. This is actually the case for a normal port, in which the approach to the throat is along a parallel-sided passage. If, however, the throat were a sudden constriction in the passage, the converging gas flow as it approached the

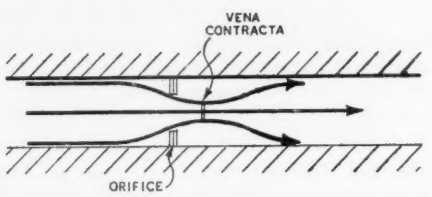


FIG. 1—Flow of gas through an orifice in a pipe

constriction would continue for some distance beyond the narrowest point. The minimum cross-section of the stream, the *vena contracta*, occurs at about half a throat diameter downstream from the throat, as shown in Fig. 1. The expression V_T in the above theory would then apply to the velocity in the *vena contracta* and not to the velocity in the throat, and ΔP would be the pressure difference between the uptake and the *vena contracta*. A gas port constructed in this way would have the advantage of giving the highest gas velocity slightly inside the furnace, but it would present serious practical difficulties. The constriction would have to be very short, otherwise the *vena contracta* would occur inside it. Such a port would very quickly burn away.

In the ports tested the flow of gas in approaching the throat has been in nearly parallel streams, and no *vena contracta* has been observed.

The normal gas port is equivalent to a nozzle giving parallel flow in the throat and discharging into a chamber of larger cross-sectional area, as in Fig. 2. After passing through a nozzle of this nature the gas stream will spread out and eventually fill the larger section some four diameters of the larger chamber downstream. The reduction of kinetic energy in this process will be accompanied by an increase of static pressure, although, owing to the formation of eddies, the reconversion will not be complete. It must be stressed that the values of F and K given subsequently for actual port designs for flow from port to furnace are for flow with no *vena contracta*, the highest gas velocity occurring in the throat. If a *vena contracta* forms, the values of F and K will be higher if based upon measurements in the throat instead of in the *vena contracta*, and this increase would not be due to greater frictional losses. The *vena contracta* did influence the port characteristics for the flow of outgoing waste gases, and this is discussed later.

(f) *Recovery of Static Pressure*

When the gas port is used for removing the waste gases from the furnace a reconversion of kinetic energy into potential energy is an advantage. The eddy loss may be reduced by a gradual expansion of pipe cross-section from its minimum to its maximum value. A semi-angle of about $2\frac{1}{2}°$ is best for this purpose,[7] although angles of up to 5° could be used with little additional loss. If a recovery of static pressure does take place it will have the effect of reducing the value of ΔP for a given gas flow. The value of F in such a flow system can then be less than unity if the recovery of pressure exceeds the frictional pressure loss ; this is the reason for the low values of F shown in Fig. 12 for certain designs of port.

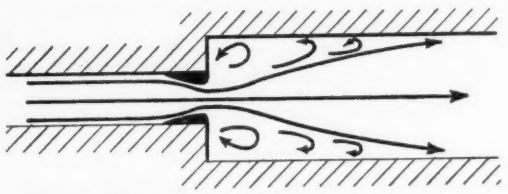


FIG. 2—Flow of gas through a nozzle in a pipe

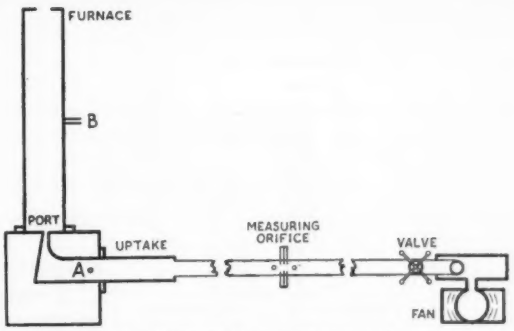


FIG. 3—General layout of apparatus

EXPERIMENTAL PROCEDURE

(a) *Apparatus*

The *K* furnace at the Templeborough works of Messrs. Steel, Peech and Tozer, Ltd., was chosen as a basis for the experiments. One-twelfth-scale models of the gas uptake and the port were constructed in the Physics Laboratories of the British Iron and Steel Research Association, the models being built on to a rectangular box which approximated to the furnace chamber without the charging doors. The experiments were carried out by blowing air through the model at such velocities that the same range of Reynolds numbers in the model port throat could be covered as occur in the full-scale furnace.

To obtain Reynolds similarity in the model a pressure drop between uptake and port of up to 5 in. W.G. was necessary. This was still small enough compared with the total gas pressure for the flow to be considered as taking place at constant density. Under these conditions the relation given in equation (5) allows the behaviour of gas flow in the prototype to be deduced from that in the model. The general layout of the apparatus is shown in Fig. 3, and details of the port construction are given in Fig. 4.

The port was built up from wooden blocks, Perspex sheet, and modelling clay, to facilitate rapid variations of design. In order further to speed the progress of the experiment, two port blocks and furnace chambers were constructed, but for accurate definition of the minimum cross-section of the port throat a single throat block was used throughout the experiments. For convenience in connecting the models to the fan the

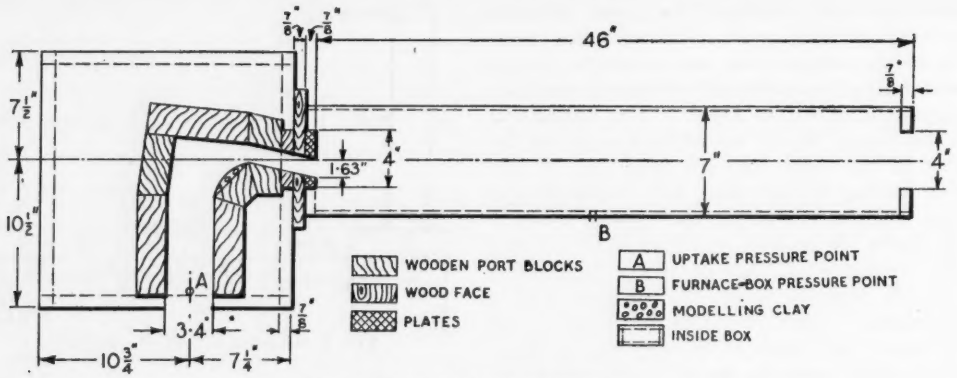


FIG 4—Details of port construction

furnace was built on its side; this does not influence the flow, as there are no differences of temperature to produce buoyancy effects. For determination of the resistances of the ports to flow in both directions it was possible to disconnect the uptake from the air supply and to connect it to the other end of the furnace.

In the full-scale furnace, R may reach a value as high as 120,000. This is based upon a kinematic viscosity of 2·77 stokes (cm.^{2} sec.$^{-1}$) or 2·981 $\times$ 10^{-3} foot-pound-second units (ft.^{2} sec.$^{-1}$) at a temperature of 1350° C. The characteristic length of the port is taken as four times the cross-sectional area divided by the perimeter. The model experiments were carried out over a range of Reynolds numbers from 25,000 to 135,000.

(b) *Flow and Pressure Measurements*

The volume of air flowing through the system was measured by means of a standard orifice meter constructed in accordance with B.S.S. 1042/1943. Simultaneous readings of this quantity, together with the pressure drop between the uptake and the furnace centre, were determined.

Values of F were plotted against values of Reynolds number in the throat, and a typical series of results for one of the ports tested is shown in Fig. 5. It is seen that F is very nearly constant for Reynolds numbers above 80,000. The apparatus for two of the ports was built up twice in order to determine whether errors of construction greatly influenced the result; Fig. 5 is made up of two such sets of results and in Fig. 6 the mean values of F for Reynolds numbers above 90,000

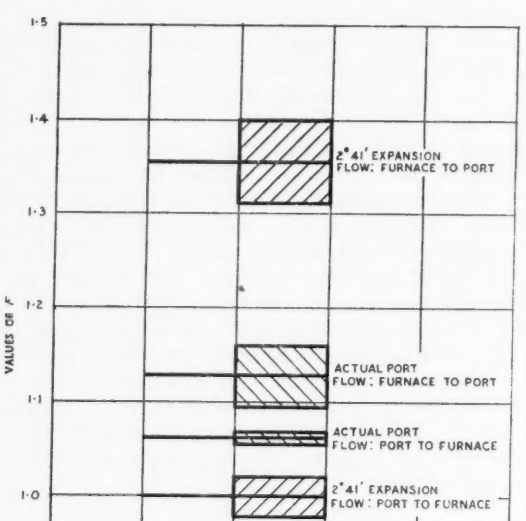

VALUES OF F
2°41' EXPANSION FLOW: FURNACE TO PORT
ACTUAL PORT FLOW: FURNACE TO PORT
ACTUAL PORT FLOW: PORT TO FURNACE
2°41' EXPANSION FLOW: PORT TO FURNACE

FIG. 6—Standard deviation of experimental F values for model gas ports

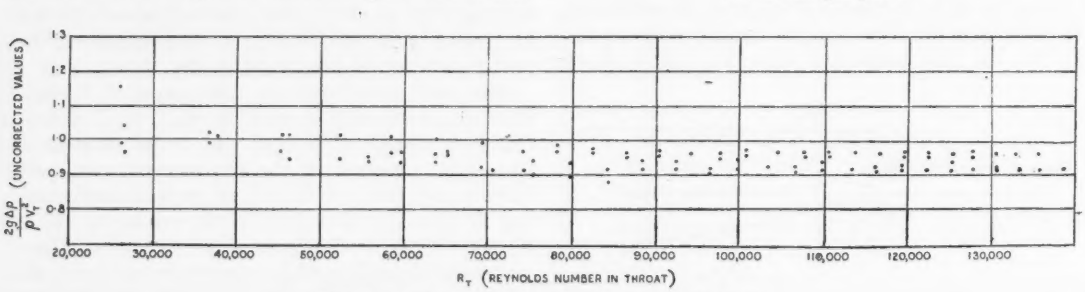


FIG. 5—Experimental results for model gas port

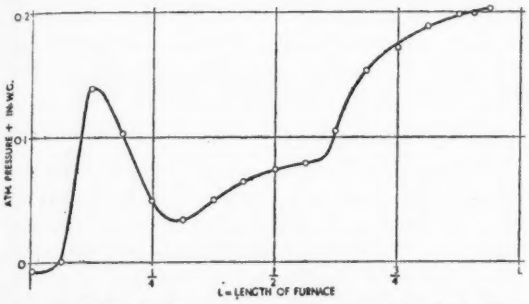


FIG. 7—Distribution of pressure throughout furnace model (pressure taken at walls). Reynolds number in throat, 107,000

are given, together with the standard deviations of the results for the two different port designs shown in Fig. 8.

(c) *Pressure Corrections*

The pressure inside the furnace chamber was measured midway between the ends of the furnace and a correction was necessary to give the true drop in pressure between uptake and throat, owing to recovery of static head, as discussed earlier in the present paper. The distribution of pressure in the model is shown in Fig. 7. The region of low pressure at about one-quarter of the furnace length from the port is caused by the ingoing stream of air being directed towards the bath and causing a local increase in velocity near the surface. The shape of the curve would be different in the full-scale furnace owing to the presence of air ports, charging doors, buoyancy of the burning gases, and the presence of outgoing ports.

RESULTS

(a) *Values of* K *for Model of Actual Port*

The value of K for the parallel port of the actual furnace shown in Figs. 8*a* and 8*b* is 1·16. This port therefore converts potential energy into kinetic energy with a loss of 16% of the energy by friction. No change of design can therefore improve the efficiency of energy conversion by more than 16%, which is equivalent to an increase of approximately 7·5% in the gas velocity.

(b) *Port with Venturi Contraction*

For the design shown in Figs. 8*c* and 8*d*, having a venturi contraction of 2° 41′ semi-angle between uptake and throat, K equals 1·09. This therefore represents a reduction of 7% in friction loss as compared with the parallel-sided port. The mean slope of the port was made the same as that of the parallel-sided port, the value being 1 in 5·33. In the original port and in all the modifications tested, the inside of the bend between uptake and throat was rounded, the radius being 60% of the uptake width.

With the existing furnace layout an expansion of much greater than 2½° is not possible. The

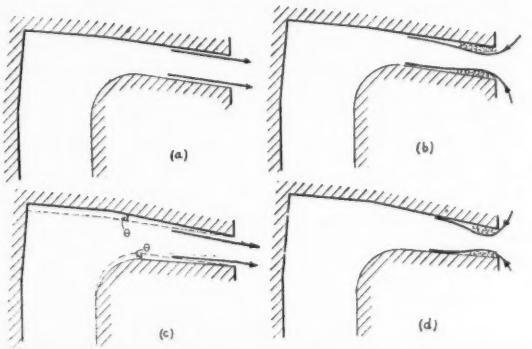


FIG. 8—Experimental gas-port models

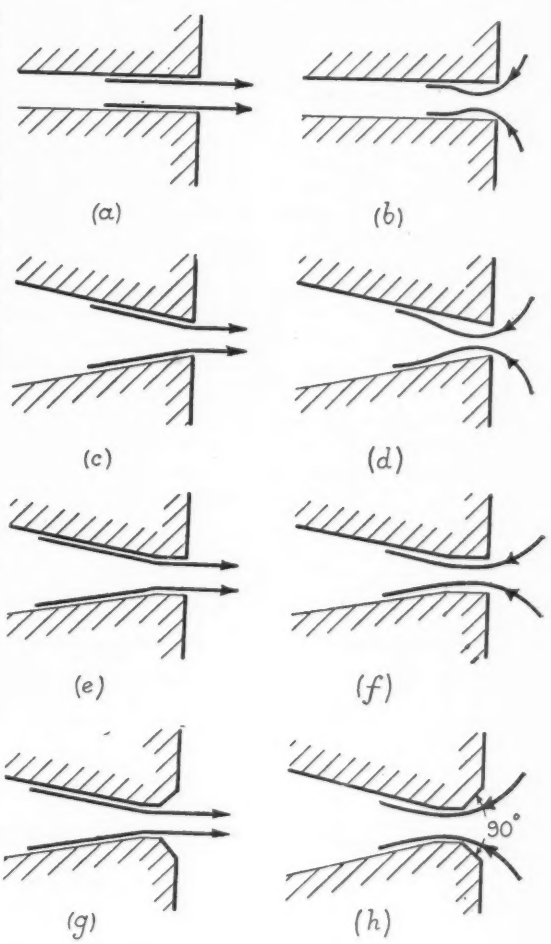


FIG. 9—Diagrammatic flow in gas ports

addition of a parallel throat between the venturi contraction and the furnace of from $D/2$ to D in length, where D is the throat width, had no effect on the resistance; this port is shown in Figs. 9*e* and 9*f*.

(c) *Port Resistances to Outgoing Gases*

In the open-hearth furnace the gas port has to play its part in removing the products of combustion from the furnace chamber. It is desirable that any modification of design leading to a reduction in ingoing resistance shall not be attained at the expense of too great an increase in the resistance to the outgoing waste gases.

Experiments showed that the port design having a 2° 41′ contraction angle between uptake and throat gives an F value of 1·36, compared to the value of 1·13 for the parallel-sided port. The addition of a parallel throat between the contraction and the furnace reduced this increased resistance to a value approximately that of the normal port with completely parallel sides.

(d) *Throat Modifications*

The effect of adding a 90° funnel between furnace and port was tried. This is shown in Figs. 9*g* and 9*h*. It was found that the ingoing resistance was entirely unaffected, but the outgoing resistance was reduced considerably, F being 0·63. A similar funnel of 120° had likewise no effect on the ingoing resistance and had a slightly higher resistance than the previous port with a 90° funnel, the F factor being in this case 0·71. The funnel lengths were $D/4$ in both cases.

The possibility of using these latter designs in practice is doubtful, as the hot gases would soon smooth off the brickwork and give the effect of a rounded trumpet instead of a sharp cone unless, of course, very effective water-cooling could be employed. While such a trumpet would aid the outgoing gas flow it would be completely useless for introducing gas to the furnace. A recovery of static head would occur in the port throat and thus reduce the entry velocity of the gas to the furnace chamber. When the port design shown in Figs. 10*d* and 10*e*, in which the length of the rounded trumpet was $D/2$, was tested, the results showed that although the F value for the outgoing flow was 0·79, the corresponding value for ingoing flow was 0·78, showing a reconversion of kinetic head into static pressure. A similar result would be expected for a port of the type shown in Figs. 9*f* and 9*g*.

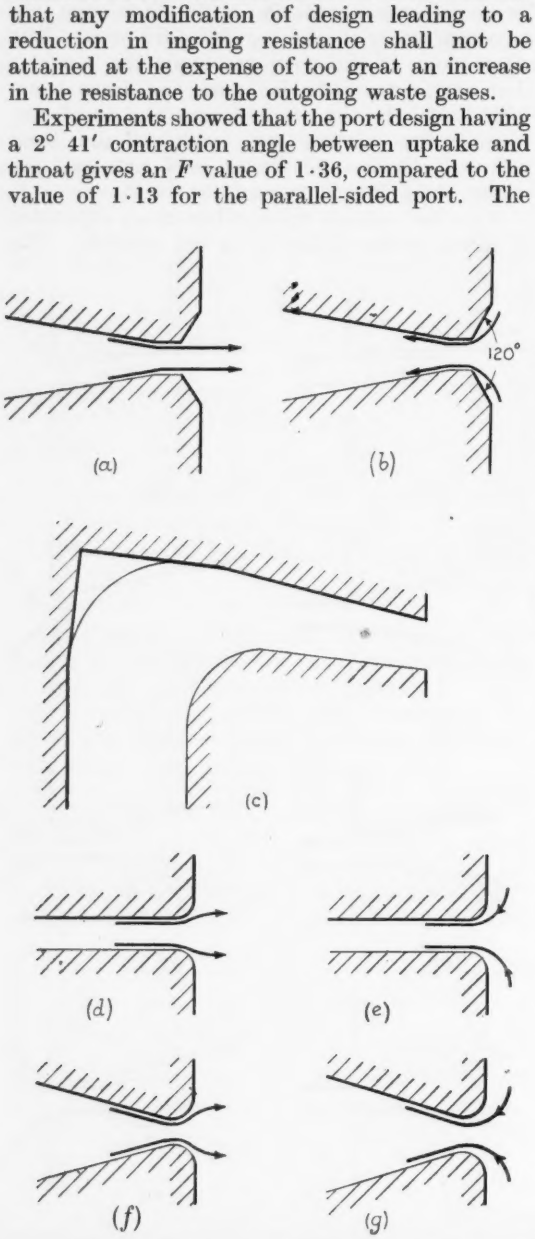


FIG. 10—Diagrammatic flow in gas ports

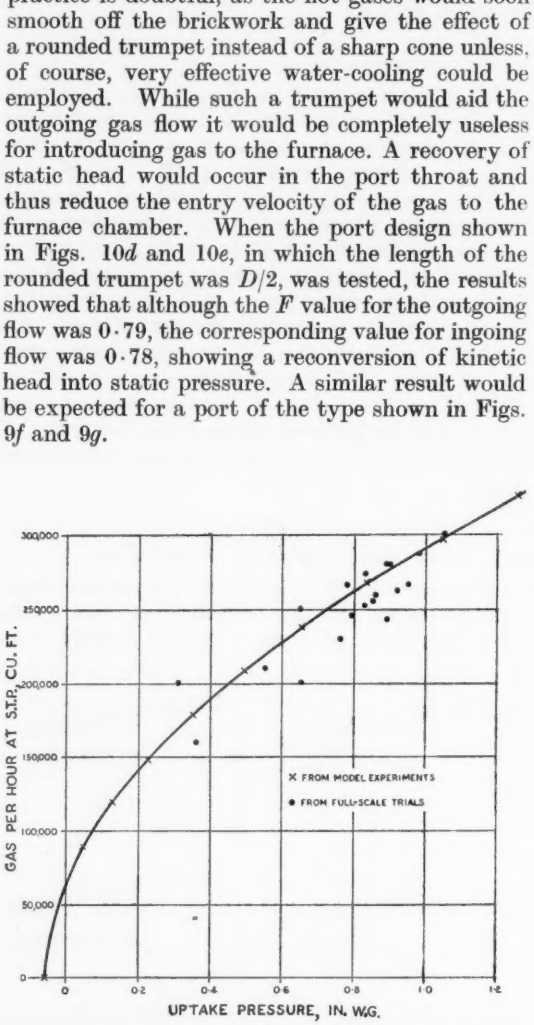


FIG. 11—Comparison of gas flow in model tests and full-scale-furnace trials

Discussion of Results

(a) *Comparison with Full-Scale Furnace*

A comparison of the behaviour of the model and the full-scale furnace is shown in Fig. 11, where the full line represents the predicted relation between uptake pressure and gas flow, and the points on either side of it are obtained from observations on the full-scale furnace.[8] The general agreement between the two shows clearly the validity of the model technique. The graph does not pass through the origin of co-ordinates because the pressure due to buoyancy of the hot gas in the original system has been subtracted from the total pressure drop calculated from the model results to give the uptake pressure as measured in the full-scale furnace.

The value of K measured in the model is 1·16, whereas that deduced for the full-scale furnace from Fig. 11 is about 1·24. The value of K for the full-scale furnace is thus about 8% higher than for the model, and this is probably due to a smaller relative roughness in the latter.

(b) *Estimation of Pressure Losses in Port with Parallel Sides*

The values of F and K for the full-scale furnace may be estimated as follows.

The process of eddy formation may be conveniently divided into two parts. The first is caused by wall friction and exists in any duct at high rates of flow, and the other is caused by the sudden change of the mean gas direction which occurs at the bend. The proportion of the kinetic energy lost due to each cause may be estimated separately, although eddies formed by both processes will occur in the region of the bend.

Experiments on right-angled bends in pipes of circular cross-section have shown that when the mean radius of the bend is 1·25 times the diameter, then some 53% of the dynamic head before the bend is destroyed in passing round the bend.[9] In the port under consideration the gas is flowing from a square uptake to a throat of **D**-shaped section, but we may take the above figure to be a rough indication of the loss of dynamic head due to the bend.

The throat area is 0·3 times the uptake area, so that the dynamic head destroyed is $0{\cdot}53 \times (0{\cdot}3)^2$ times the dynamic head in the throat, or approximately 0·05 times the throat dynamic head.

The pressure drops in the uptake and throat may be estimated from a formula of the type :

$$\Delta p \times f \frac{LC}{4A}, \qquad \text{.....(9)}$$

where L and C are the length and perimeter of the duct, respectively, and A is its cross-sectional area. The value of f depends on the Reynolds number and upon the roughness of the duct walls, but may be taken as about 0·035 for furnace brickwork[6] (*loc. cit.*, pp. 154–155). Inserting values for L, C, and A we find that for the uptake between the point at which the pressure is measured and the bend, 0·14 dynamic heads will be destroyed, or 0·04 throat dynamic heads. Similarly the sloping part of the port will destroy 0·12 throat dynamic heads, so that altogether $(0{\cdot}05 + 0{\cdot}04 + 0{\cdot}12) = 0{\cdot}21$ throat dynamic heads will be destroyed.

The estimated value of F is therefore 1·21, and the value of K is 1·34. This estimated value is some 8% higher than the value found from the furnace trials and may be attributed to the uncertainty in the value which should be given to f owing to uncertainty in the roughness of the brickwork. The estimation is, however, of interest in showing that in the normal port the greatest single contribution to the frictional pressure drop of gas entering the furnace occurs in the parallel section of the port.

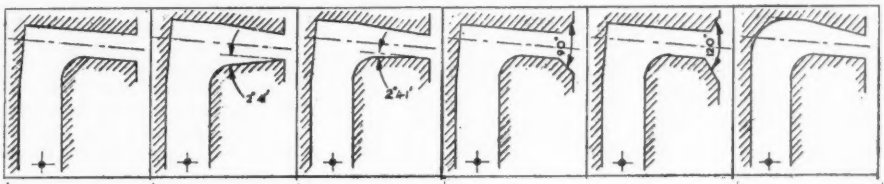
PORT DESIGN						
Flow out of Furnace :						
F	1·13	1·36	1·19	0·63	0·71	1·22
Flow into Furnace :						
K	1·16	1·09	1·09	1·09	1·09	1·06
F	1·06	1·00	1·00	1·00	1·00	0·97
Ω	0·93	0·95	0·96	0·95	0·96	0·97

Fig. 12—Experimental results for model gas ports

The values of F and K in the various model tests are strictly comparable, but it appears that owing to greater friction losses in an actual furnace the tabulated values should be increased by about 8%.

(*c*) *Port Modifications*

The reduction of resistance for ingoing gas flow which results from the venturi contraction is due to two causes. The first is a reduction of wall resistance in the converging portion of the port. The converging walls are very effective in preventing eddies, and a few degrees of convergence will give the effect of almost suppressing eddy formation[7] (*loc. cit.*, p. 51). Secondly, a slight reduction of resistance is probably due to the less sudden contraction of the stream which occurs when the cross-section of the flow changes from that of the uptake to that of the port.

The increase in outgoing resistance in such a port is due to the *vena contracta*. It is smaller in comparison with the cross-sectional area of the port at the point at which it occurs than in the parallel-sided port. A rather greater loss of energy due to eddy formation thus occurs than for the outgoing flow in the parallel-sided port. The results show that the addition of a short parallel throat brings the resistance back to the value for a completely parallel port.

Within the length of the venturi tested, which is some three times the throat width, the influence of the expansion in causing a recovery of static head before the bend is negligible, and most of the outgoing resistance is due to the bend. The reduction in outgoing resistance when the outside of the bend between port and downtake is rounded may be due to recovery of static head being able to continue to some extent round the bend.

Values of K and F for all the ports tested are shown in Fig. 12.

Conclusions

The above model results show that for a port of the type tested having a well-rounded bend, no improvements can add more than about 7% to the ingoing velocity. For a given pressure drop the best modifications have added some 4% to this gas velocity, which, for the model, will only be 3% short of its theoretical maximum. In an actual furnace, owing to greater wall-friction losses, the velocity would be within about 7% of its theoretical maximum for a design of this sort.

From a practical point of view the port shown in Fig. 9*e* would appear to be the best. With a parallel throat having a length approximately equal to its width the characteristics of the port would not be adversely affected if the throat were to burn back say half the length of the straight portion.

It appears then that if any substantial increase in the entry velocity of the gas is required it can only be brought about by an increase in the pressure drop across the port. This course may present practical difficulties, but these must be solved if effectively greater velocities are to be achieved.

Appendix—*Coefficient of Discharge*

It is useful to give an alternative method of expressing the efficiency of a port. Instead of finding the ratio of the available potential energy to the kinetic energy developed, which is expressed by K, we may find the ratio of the quantity of gas flowing per second to that which would flow through a frictionless port of the same size and shape for the same total pressure drop. This ratio is called the coefficient of discharge and is denoted by Ω. For a perfect port $2g\Delta P/\rho(V_T^2 - V_U^2) = 1$, but in an actual port $2g\Delta P/\rho(W_T^2 - W_U^2) = K$, where W_T and W_U are the lower velocities developed in this case. Thus $K = (V_T^2 - V_U^2)/(W_T^2 - W_U^2)$. But $V_U/V_T = W_U/W_T = C$ say, therefore $K = V_T^2/W_T^2 = 1/\Omega^2$ or $\Omega = 1/\sqrt{K}$.

Values of Ω are tabulated in Fig. 12, together with the values of F and K.

Acknowledgments

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The Evolution and Absorption of Hydrogen by Steel in Industrial Reheating Furnaces*

By J. Cameron, Ph.D., B.Sc., A.R.I.C.†

SYNOPSIS

The influence of reheating-furnace atmospheres on the hydrogen content of steel ingots is discussed from the theoretical standpoint.

The hydrogen content of furnace atmospheres is calculated for a variety of fuels and for different conditions of humidity and combustion, and it is shown that it is negligible for all practical conditions. The possibility of local concentrations of hydrogen occurring due to the action of water vapour on iron is considered, and it is shown that such local concentrations are not likely to exceed 2% of hydrogen.

The effect of such small concentrations of hydrogen on the equilibrium hydrogen content of the steel and also on the removal or absorption from steels not in equilibrium is examined. It is shown that due to the large masses of steel and the relatively short soaking times met with in works practice, the hydrogen content of reheating-furnace atmospheres has practically no effect on the hydrogen content of the steel ingot.

Of all the impurities with which steel becomes contaminated during the process of manufacture, hydrogen is probably the one which is most troublesome to the steelmaker. Although it is normally present in steel in concentrations of the order of 10^{-3} to 10^{-4} per cent, it is responsible for the rejection of thousands of pounds worth of steel annually. The chief trouble is caused by hair-line cracks in wrought steels, and it is fairly well established that such cracks are due to or at least associated with the presence of hydrogen in steel, but the exact mechanism of crack formation has not been fully elucidated.

By far the greatest proportion of the hydrogen present in steel is undoubtedly introduced during the period in which the steel is in the molten state; the main sources of contamination being (*a*) scrap covered with hydrated rust, (*b*) moist furnace additions of fluxes, (*c*) atmospheric humidity, and (*d*) ferro-alloys which frequently contain large quantities of hydrogen. From time to time the question has been raised of the possibility of hydrogen absorption occurring during the reheating of steel prior to rolling. Whether hydrogen is absorbed by the steel from the furnace atmosphere or whether hydrogen is evolved by the steel into the atmosphere depends on a number of factors, and it is the intention of the present paper to discuss the problem in some detail. The discussion is based on the experimental results of Andrew, Lee, Mallik, and Quarrell described in their paper, "The Removal of Hydrogen from Steel."[1] It is based particularly on their results for the removal of hydrogen from the 3%-Cr–Mo steel *HD* at 1100° C., both in a hydrogen-free atmosphere and in atmospheres containing small amounts of hydrogen. It is realized that the temperature of 1100° C. used throughout in that work is considerably lower than that normally met with in industrial reheating furnaces, and that, moreover, the calculations will be strictly true only for the 3%-Cr–Mo steel *HD*. The workers mentioned have shown, however, that the diffusion constant is practically the same for all the steels examined, and it is felt that the conclusions arrived at will hold for steels other than the 3%-Cr–Mo steel *HD*. Also, the manner in which the diffusion constant varies with temperature is known, and it is considered that the conclusions reached for a temperature of 1100° C. will be valid for slightly higher temperatures.

Whether hydrogen is evolved from or absorbed by a steel in a reheating furnace depends on three factors, *viz.*, (*a*) the solubility of hydrogen in the steel, (*b*) the partial pressure of hydrogen in the furnace atmosphere, and (*c*) the actual hydrogen content of the steel. We shall consider these three factors in turn.

The Solubility of Hydrogen

The three factors which influence the solubility of hydrogen in steel are temperature, pressure,

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† Clyde Alloy Steel Co., Ltd., Motherwell.

and the composition of the steel. The influence of temperature does not concern us in the present discussion since we are only considering the state of affairs at one temperature, namely, 1100° C.

It has been shown by Sieverts[2] that the solubility of hydrogen in pure iron is proportional to the square root of the hydrogen pressure. It is therefore only necessary to know the solubility at any given pressure to enable us to calculate it at any other. The basis of reference usually used is the solubility in c.c./100 g. at a hydrogen pressure of 1 atm. In their experiments on the isothermal removal of hydrogen from steel, Andrew and his co-workers soaked specimens of $1\frac{3}{4}$ in. dia. and 2 in. long for 24 hr. at 1100° C. in hydrogen at a pressure of 1 atm. The hydrogen content of the specimens after quenching was found to be between 7·65 and 7·79 c.c./100 g., for the steel *HD*. If we assume that 24 hr. is sufficient time for the specimens to come to equilibrium with the hydrogen atmosphere, we can take these figures as representing the solubility of hydrogen in steel *HD* under a hydrogen pressure of 1 atm. That the assumption that 24 hr. is sufficient time to attain equilibrium is justified can be shown as follows.

Andrew and his co-workers showed, both experimentally and theoretically, that the rate of evolution of hydrogen from steel *HD* at 1100° C. follows a law of the form :

$$y = \frac{k}{1+a}\,(1 + ae^{-nt}),$$

where y is the hydrogen content of the steel at time t, and k, a, and n are constants. From a study of the theoretical derivation of the law it is apparent that it should also be applicable to the absorption by the steel of hydrogen from the atmosphere ; in this case the limiting concentration y_1 being the solubility of hydrogen in the steel at the hydrogen pressure of the experiment, namely, 1 atm. It can also be shown that $y_1 = k/(1+a)$, and $a = (y_0 - y_1)/y_1$, y_0 being the hydrogen content of the steel at time zero. Since in the present case $y_0 = 0$ and hence $a = -1$, the equation can be rewritten :

$$y = y_1\,(1 - e^{-nt})$$

Since at 1100° C., n has a value of 1·2, the hydrogen content of the steel after 24 hr. soaking is given by the expression :

$$y_{24} = y_1\,(1 - e^{-28\cdot8})$$

For all practical purposes therefore, the limiting concentration is reached in 24 hr.

That the influence of the composition of the steel on the hydrogen solubility is of minor importance is shown by the fact that for the five steels examined by Andrew *et alia*, the values obtained were 5·78, 6·03, 6·40, 6·46, and 7·7 c.c./100 g., and also by the fact that these values differ little from the value of 6·6 c.c./100 g. given by Sieverts[2] and 7·75 c.c./100 g. given by Luckmeyer and Hasse[3] for pure iron. Throughout this discussion which, as has been previously explained, is based on the results for steel *HD*, the value of the hydrogen solubility under 1 atm. pressure will be taken as 7·8 c.c./100 g.

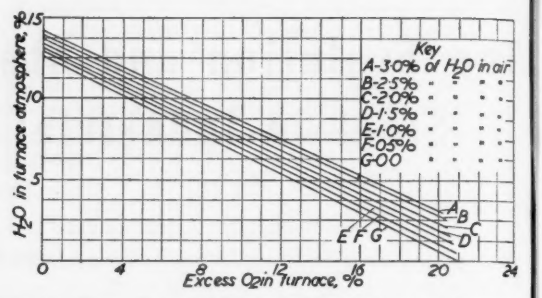

FIG. 1—Percentage of H_2O in atmosphere of furnace fired by producer gas (of composition : CO, 24% ; CO_2, 6% ; CH_4, 3% ; H_2, 13% ; H_2O, 6%)

The Furnace Atmosphere

In discussing the possible absorption or evolution of hydrogen from steel in industrial reheating furnaces it is essential to know the hydrogen content, and as we shall see later, also the water content of the furnace atmosphere. The factors which affect the composition of a furnace atmosphere are (*a*) the composition of the fuel used, (*b*) the humidity of the air, and (*c*) the state of combustion. The third factor may be quantitatively stated as the percentage of excess oxygen present after combustion when under oxidizing conditions, or the ratio CO/CO_2 under reducing conditions. When the three factors above are all known, the composition of the furnace atmosphere can be calculated from the volume relationships in the chemical equations representing the combustion of the various constituents of the fuel. Figure 1 shows the variation with air humidity and percentage of excess oxygen, of the water content of the atmosphere of a furnace fired by

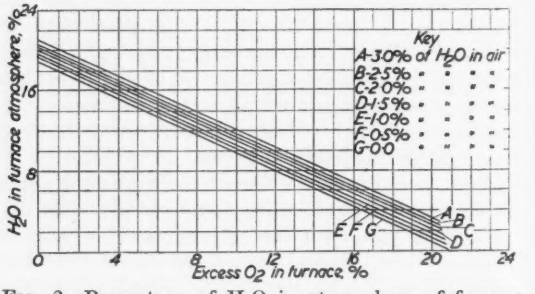

FIG. 2—Percentage of H_2O in atmosphere of furnace fired by blue gas (of composition : CO, 43% ; CO_2, 4% ; CH_4, 1% ; H_2, 48% ; H_2O, 4%)

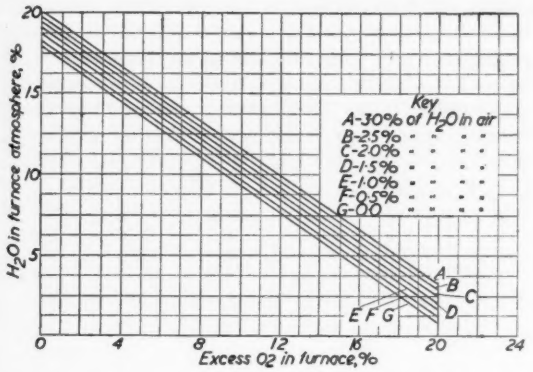

FIG. 3—Percentage of H_2O in atmosphere of furnace fired by carburetted water gas (of composition : CO, 30·5% ; CO_2, 6% ; CH_4, 14% ; C_2H_4, 7% ; H_2, 37% ; H_2O, 4%)

producer gas of the following typical composition :

CO, %	CO_2, %	CH_4, %	H_2, %	H_2O, %
24	6	3	13	6

Figures 2, 3, and 4 are for furnaces fired by blue gas, carburetted water gas, and town's gas, respectively, while Fig. 5 is for a furnace fired by coal. The compositions of these fuels are as follows :

	C, %	CO %	CO_2, %	CH_4, %	C_2H_4, %	H_2, %	H_2O %	Ash, %
Blue gas		43	4	1		48	4	
Carburetted water gas		30·5	6	14	7	37	4	
Town's gas ...		16	4	18	2	54		
Coal	80					5	5	10

When conditions are such that combustion of the fuel is incomplete the calculation of the hydrogen and water contents of the furnace atmosphere is much more complex and it is not possible to draw simple diagrams of the type used when an excess of oxygen is present. Table I, however, shows the values calculated for each of the above fuels under mildly reducing conditions, characterized by a CO/CO_2 ratio of 1/10 in the burnt gas, and also under the strongly reducing conditions described by a CO/CO_2 ratio of 1/2. In addition, each one is calculated for moist summer and dry winter weather with 3·1% of H_2O and 0·3% of H_2O, respectively, in the atmosphere.

We are now in a position to discuss the hydrogen content of the general furnace atmosphere. Under oxidizing conditions we may take 3% of excess oxygen as being fairly typical of industrial conditions. We can then calculate the hydrogen content of the gas from the equilibrium constant of the reversible reaction :

$$2H_2 + O_2 \rightleftharpoons 2H_2O$$

Using partial pressure in atmosphere, we have :

$$\frac{p^2H_2 \times pO_2}{p^2H_2O} = K.$$

Now at 1100° C., $\log K = -13{\cdot}012$. Therefore :

$$\frac{p^2H_2}{p^2H_2O} = \frac{K}{pO^2} = \frac{10^{-13{\cdot}012}}{3 \times 10^{-2}} = 3{\cdot}24 \times 10^{-13}$$

$$\therefore \frac{pH_2}{pH_2O} = 1{\cdot}8 \times 10^{-6}.$$

We see, therefore, that under oxidizing conditions the hydrogen content of the general furnace atmosphere is extremely small and may be neglected. Under reducing conditions the hydrogen content of the atmosphere may be considerable. Table I shows that under mildly reducing conditions it may be up to 1·1%, whilst under strongly reducing conditions hydrogen contents of up to 5% are possible. The conditions described as strongly reducing are not very likely to occur in industrial practice, and even those described as mildly reducing are extremely rare, since most steelmakers prefer to work with a slight excess of air in order to assure complete combustion of the fuel.

There is, however, another factor which must be taken into consideration. This is the presence of a local concentration of hydrogen at the surface of the steel, due to reactions such as :

(*a*) $Fe + H_2O \rightleftharpoons H_2 + FeO$
(*b*) $3\,Fe + 4H_2O \rightleftharpoons 4H_2 + Fe_3O_4$
(*c*) $3\,FeO + H_2O \rightleftharpoons H_2 + Fe_3O_4$

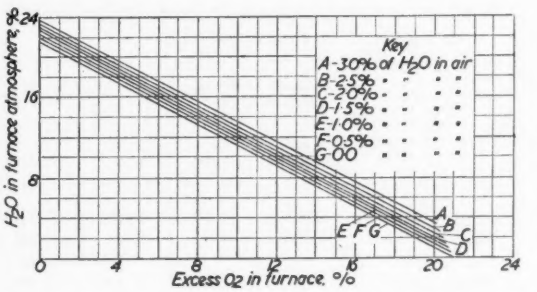

FIG. 4—Percentage of H_2O in atmosphere of furnace fired by town's gas (of composition : CO, 16% ; CO_2, 4% ; CH_4, 18% ; C_2H_4, 2% ; H_2, 54%)

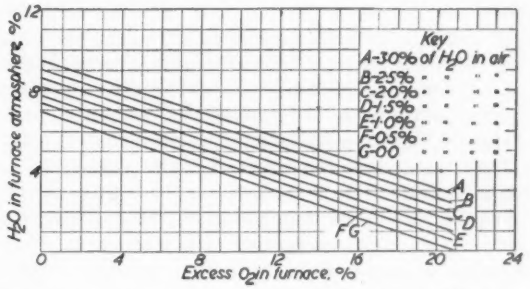

FIG. 5—Percentage of H_2O in atmosphere of furnace fired by coal (of composition : C, 80% ; H_2, 5% ; H_2O, 5% ; ash, 10%)

TABLE I—*Hydrogen in Furnace Atmosphere*

Fuel	Mildly Reducing Conditions				Strongly Reducing Conditions			
	Moist Summer		Dry Winter		Moist Summer		Dry Winter	
	H_2O, %	H_2, %	H_2O, %	H_2, %	H_2O, %	H_2, %	H_2O, %	H_2, %
Producer gas	14·2	0·7	12·7	0·6	13·3	2·9	12·3	2·7
Blue gas	21·1	1·0	19·4	0·9	20·3	4·7	19·2	4·3
Carburetted water gas	20·5	0·9	17·2	0·8	20·8	4·7	18·3	4·1
Town's gas	23·5	1·1	21·2	1·0	22·0	5·0	20·8	4·7
Coal	9·7	0·4	7·4	0·3	8·9	2·0	6·7	1·5

Although it is not possible to calculate the actual value of hydrogen which this local concentration will have, it is possible to calculate the maximum or equilibrium value which it can attain if the steel is to remain scaled. In order to do this we must first discuss the mechanism of scaling of steel by oxidizing atmospheres.

It is generally accepted that the scale formed on steel at high temperatures consists of three layers ; *viz.*, a thin outer layer of Fe_2O_3, an intermediate layer of Fe_3O_4 and an inner layer of wüstite. According to Schenck[4] a mixture of steam and hydrogen in equilibrium with iron and wüstite contains 57% of H_2 ($H_2/H_2O = 57/43$) ; the mixture in equilibrium with wüstite and Fe_3O_4 contains 10% of H_2 (*i.e.*, $H_2/H_2O = 1/9$) ; whilst for the equilibrium between Fe_3O_4 and Fe_2O_3 the hydrogen content is too small to measure. These equilibrium conditions are not affected by changes in pressure or by the presence of other gases, since the equilibrium constants of the reactions concerned involve only the ratio of hydrogen to water vapour :

$$Fe + H_2O \rightleftharpoons H_2 + FeO, \therefore K_1 = \frac{pH_2}{pH_2O}.$$

$$3\,Fe + 4H_2O \rightleftharpoons 4H_2 + Fe_3O_4, \therefore K_2 = \frac{P^4H_2}{p^4H_2O} = \left(\frac{pH_2}{pH_2O}\right)^4.$$

$$3\,FeO + H_2O \rightleftharpoons H_2 + Fe_3O_4, \therefore K_3 = \frac{pH_2}{pH_2O}.$$

We see, therefore, that in order to determine the probable hydrogen content produced at the steel surface as a result of the scaling action of water, it becomes of the utmost importance to know the exact mechanism of scaling by water vapour. That is, whether the steam reacts at the Fe_2O_3–Fe_3O_4 interface, the Fe_3O_4–wüstite interface, or the wüstite–iron interface. According to the most widely accepted theory of the mechanism of scaling, the counter diffusion theory of Pfeil,[5] oxidation occurs by a diffusion of iron through the wüstite and a diffusion of oxygen, probably in the form of Fe_3O_4, also through the wüstite towards the base metal. It would therefore appear that the equilibrium between steel, wüstite, hydrogen and steam does not enter into the question at all. It appears very probable that the mechanism of scaling by steam consists of the steam reacting with some of the FeO in the wüstite to form Fe_3O_4 according to the equation :

$$3\,FeO + H_2O \rightleftharpoons Fe_3O_4 + H_2.$$

The Fe_3O_4 diffuses inwards through the wüstite, meeting Fe which is diffusing outwards and reacting with it according to the following equation :

$$Fe_3O_4 + Fe \rightleftharpoons 4\,FeO$$

If that is the true mechanism of scaling then the important equilibrium is the one between wüstite, Fe_3O_4, H_2, and H_2O, which occurs when the ratio $pH_2/pH_2O = 1/9$ at 1100° C. It follows, therefore, that the maximum hydrogen concentration which can be built up at the surface of scaled steel is one tenth of the $H_2 + H_2O$ content of the furnace atmosphere.

Table II shows this maximum hydrogen concentration for industrial furnaces fired by various fuels, under oxidizing and mildly reducing conditions. Oxidizing conditions are characterized by 3% of excess oxygen in the furnace atmosphere.

Under conditions described as strongly reducing, the general hydrogen content of the atmosphere already exceeds 1/10 of the $H_2 + H_2O$ content of

TABLE II—*Hydrogen Concentration at the Surface of Scaled Steel in Various Furnace Atmospheres at 1100°*

Fuel	H_2 under Oxidizing Conditions, %		H_2 under Mildly Reducing Conditions	
	Moist Summer	Dry Winter	Moist Summer	Dry Winter
Producer gas ...	1·26	1·10	1·49	1·33
Blue gas ...	1·86	1·66	2·21	2·03
Carburetted water gas ...	1·74	1·57	2·14	1·80
Town's gas ...	2·07	1·89	2·45	2·22
Coal	0·86	0·63	1·01	0·77

the gas. Under these conditions, therefore, there can be no Fe_3O_4 layer on the scale which will in this case consist of wüstite only. Moreover, the concentration of Fe_3O_4 in the wüstite will not increase beyond the value in equilibrium with the ratio of H_2/H_2O in the furnace atmosphere. Diffusion of Fe_3O_4 from the surface inwards will still occur, however, the concentration of Fe_3O_4 in the inner layer being continually reduced due to the reaction :

$$Fe_3O_4 + Fe \rightleftharpoons 4\,FeO$$

The steel will therefore be scaled even under these conditions, but no local concentration of hydrogen will occur.

From the foregoing discussion we see that in an industrial reheating furnace there is the possibility of a local concentration of hydrogen of up to 2% at the surface of the steel. The effect of this concentration of hydrogen can be discussed under two headings :

(*a*) The effect on equilibrium conditions.

(*b*) The effect on the rate of evolution or absorption of hydrogen.

Equilibrium Conditions

It is a well-known fact that the solubility of hydrogen in steel is proportional to the square root of the pressure. This law will hold even in the presence of other gases, the solubility then becoming proportional to the square root of the partial pressure of hydrogen in the gas. Therefore, in order to calculate the hydrogen content of a steel in equilibrium with a given atmosphere it is sufficient to know the hydrogen content of the atmosphere and the solubility of hydrogen in the steel at that temperature under a pressure of 1 atm. of hydrogen. For the steel *HD* the solubility at 1100° C. and 1 atm. pressure is 7·8 c.c./100 g. The hydrogen content in this steel in equilibrium with a given atmosphere is therefore given by the expression :

$$H = 7{\cdot}8\sqrt{pH_2}$$
$$= 0{\cdot}78\sqrt{H_2\,(\%)\text{ in atmosphere.}}$$

TABLE III—*Equilibrium Hydrogen Content of Steel* HD

Fuel	H_2 under Oxidizing Conditions, c.c./100 g.		H_2 under Mildly Reducing Conditions, c.c./100 g.	
	Moist Summer	Dry Winter	Moist Summer	Dry Winter
Producer gas ...	0·87	0·82	0·95	0·90
Blue gas ...	1·06	1·00	1·16	1·11
Carburetted water gas ...	1·03	0·98	1·14	1·05
Town's gas ...	1·12	1·07	1·22	1·16
Coal	0·72	0·62	0·78	0·68

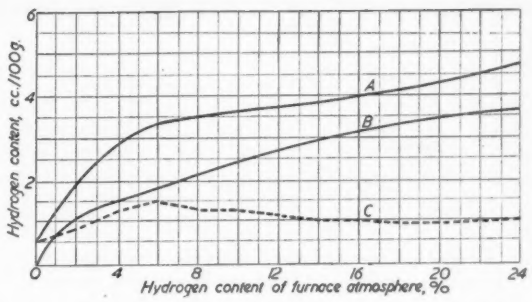

FIG. 6—Results of soaking steel *HD* for 5 hr. at 1100° C. in atmospheres of different hydrogen content

Table III shows the equilibrium concentration for steel *HD*, calculated for different industrial atmospheres under various conditions of humidity and combustion.

If the initial hydrogen content of the steel exceeds the equilibrium value for the ruling conditions, then evolution of hydrogen from the steel into the furnace atmosphere will occur. If the initial value is less than the equilibrium value, absorption of hydrogen into the steel from the furnace atmosphere will take place.

We see from Table III that when in equilibrium with the atmosphere of an industrial reheating furnace, steels should have a hydrogen content of approximately 1 c.c./100 g. It is very probable, however, that in works practice, where large masses of steel are heated for comparatively short periods of time, equilibrium conditions are never attained. The rate of approach towards equilibrium, *i.e.*, the rate of reaction, then becomes much more important than the equilibrium value itself. Andrew and his collaborators investigated the rate of evolution of hydrogen from steel at 1100° C. into atmospheres free from hydrogen, and found that it obeyed a law of the form :

$$y = \frac{K}{1+a}(1 + ae^{-nt}),$$

where y is the hydrogen content at time t, and k, a, and n are constants. They also investigated the effect of hydrogen in the furnace atmospheres on the evolution of hydrogen from steel. Curve *A*, Fig. 7, shows their results for the hydrogen content of steel *HD* after 5 hr. heating at 1100° C. in atmospheres of different hydrogen content. The steel samples initially contained between 7·65 and 7·8 c.c. of hydrogen per 100 g. in each case. Curve *B* shows the equilibrium hydrogen content as calculated from the simple solubility law. It is apparent that five hours heating has not been sufficiently long to allow the attainment of equilibrium even in the small specimens ($1\frac{3}{4}$ in. dia. × 2 in. long) used in the experiments. Andrew and co-workers also derived theoretically from

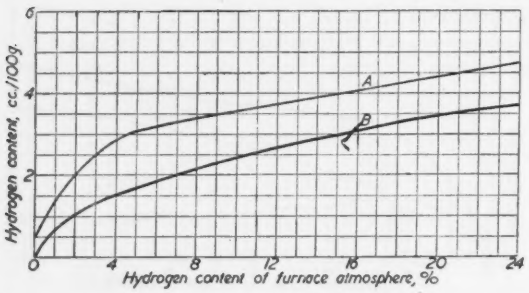

FIG. 7—Effect of hydrogen in furnace atmosphere on limiting concentration of hydrogen

Fick's law an expression of the same form as that obtained experimentally. In deriving this expression they made the assumption that there is a lower limiting concentration of hydrogen for each steel. This assumption was necessary because if the lower limit were zero, the expression obtained was an exponential form which did not give good agreement with the experimental results. We shall now examine the possibility of extending the use of their equation to the case where the atmosphere contains a percentage of hydrogen. In the derivation of the expression the only assumptions made are :

(i) That Fick's law is obeyed.
(ii) At zero time the hydrogen concentration is uniform throughout the steel.
(iii) At time infinity, the lower limiting hydrogen concentration has been reached.

We see, therefore, that there is no reason why the same type of expression should not hold for the case where the atmosphere contains hydrogen. It is to be expected, of course, that the limiting concentration will vary with the hydrogen content of the atmosphere, since in addition to the limiting concentration (whatever its causes) found when the atmosphere is hydrogen free, there is also the limit imposed by the solubility in the steel of hydrogen at the partial pressure at which it is present in the furnace atmosphere. Assuming that in an atmosphere containing hydrogen, the relation between hydrogen content and time can be expressed by an equation of the form :

$$y = \frac{K}{1+a} \quad (1 + ae^{-nt}),$$

and also that the value of n does not vary with the hydrogen concentration in the atmosphere, we can calculate from the initial and final contents the variation of the limiting hydrogen content with the hydrogen concentration in the furnace atmosphere. Figure 6 shows the limiting concentration calculated by the above method (curve A) the equilibrium concentration calculated from solubility law (curve B), and the difference between these two values (curve C). We see that the limiting concentration postulated by Andrew exceeds the solubility value by an amount which varies from 0·5 c.c./100 g. to 1·5 c.c./100 g. The amount of the difference does not show any relationship to the hydrogen content of the atmosphere. It is difficult to tell whether this difference between the limiting and solubility concentrations has any real significance or whether it is merely due to a limitation of the experimental and theoretical treatments. Whichever is the true state of affairs it does not affect the present discussion, since we know from Andrew's experiments how the limiting concentration varies with the hydrogen content of the atmosphere. In the case of other steels for which experimental results are not available it would appear to be justifiable to obtain the value of the limiting concentration by adding 1 c.c./100 g. to the equilibrium value calculated from the solubility law.

Rate of Evolution or Absorption

If we accept the validity of the above reasoning we are now in the position to draw curves showing the rate of evolution of hydrogen from steel held at 1100° C. in an atmosphere of any known hydrogen concentration. The most probable figure for the hydrogen content of a commercial ingot is 6 c.c./100 g., and we have shown that due to scaling reactions the usual hydrogen content of an industrial-furnace atmosphere is probably between 1% and 2%. We shall, therefore, derive equations for the removal of hydrogen from steel *HD* at 1100° C. in atmospheres containing 0, 1, and 2% of hydrogen, starting with an initial hydrogen content of 6 c.c./100 g. of steel. The equations will, of course, hold only for the size and shape of specimen used by Andrew *et alia* in their experiments, and will be of the form :

$$y = \frac{K}{1+a}(1 + ae^{-1\cdot 2t}).$$

Since when $t = 0$, $y = k$. Therefore $k = 6$.

Also when $t = \infty$, $y = k/(1 + a) = y_1$, where y_1 is the limiting concentration.

Therefore :

$$a = \frac{6 - y_1}{y_1}$$

The three equations required are therefore :

$y = 0{\cdot}482\ (1 + 11{\cdot}45\ \ e^{-1\cdot 2t})$; where atmosphere is hydrogen free.
$y = 1{\cdot}25\ (1 + 3{\cdot}80\ \ e^{-1\cdot 2t})$; where there is 1% of H_2 in atmosphere.
$y = 1{\cdot}90\ (1 + 2{\cdot}158\ \ e^{-1\cdot 2t})$; where there is 2% of H_2 in atmosphere.

Figure 8 is a graphical representation of the evolution of hydrogen under these three circumstances.

In addition to the evolution of hydrogen from steel, it is also of interest to consider the rate of

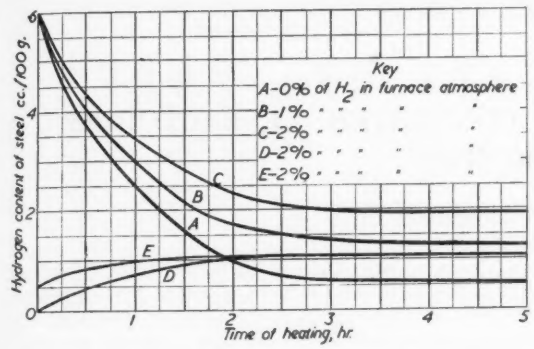

FIG. 8—Evolution and absorption curves for hydrogen in steel *HD*

absorption of hydrogen by steel which has initially a very low hydrogen content. From the theoretical standpoint there is nothing in the derivation of the equation for hydrogen removal which could prevent it from being applied also to hydrogen absorption. There is, however, the difficulty of deciding the best value for the limiting concentration ; that is, whether a value less than or greater than the value obtained from the solubility law should be taken. In the present case, in the absence of any definite guide, the actual solubility law value will be taken. Figure 8 (curves *D* and *E*) shows the rate of absorption of hydrogen by steel *HD* with initial hydrogen concentrations of 0 and 0·5 c.c./100 g., respectively, when heated at 1100° C. in an atmosphere containing 2% of H_2, the maximum value likely to occur in an industrial furnace. It should, of course, be remembered that the equations and graphs given above will only be true for the size and shape of specimen used in the experimental work, namely, $1\frac{3}{4}$ in. dia. × 2 in. long. Andrew and co-workers have shown, however, that the effect of specimen size is merely to change the value of n, and that n is proportional to the square of the diameter for specimens of the same shape. Since the constant n appears in the equation as nt it follows, therefore, that the only effect of change in specimen size is to alter the time scale of the graphs in Fig. 8. From these graphs we see that whether or not 1% or 2% of hydrogen in the furnace atmosphere has any appreciable effect on hydrogen evolution, depends on the size of the steel ingot and on the time of treatment. Now most commercial ingots are at least ten times the diameter of the specimen used in the experimental work on which these curves are based, and so the time scale should be multiplied one hundredfold. For reasonable soaking times, therefore, the conditions are represented by the portion of the curves lying close to the y axis, where the effect of small amounts of hydrogen in the atmosphere is negligible. The following figures in Table IV illustrate the small differences obtained with steel ingots originally containing 6 c.c. of hydrogen per 100 g. of steel.

TABLE IV—*Effect of Specimen Size on Hydrogen Evolution*

Dia. of Ingot, in.	Time of Soaking, hr.	Final Hydrogen Content, c.c./100 g.		
		Hydrogen in Atmosphere		
		0%	1%	2%
17·5	25	4·57	4·77	4·94
17·5	50	3·83	4·13	4·39
21	48	4·3	4·55	4·75

From the above considerations we can conclude, therefore, that in the reheating of steel ingots for rolling, the amount of hydrogen removed is probably less than 1 c.c./100 g., and also that the rate of removal is not appreciably affected by the presence in the furnace atmosphere of up to 2% of hydrogen which, as we have shown, is the maximum likely to occur in an industrial reheating furnace. We also see from the curves of Fig. 8 that where the steel is initially quite or nearly hydrogen free, reheating for the times normally used in commercial practice in a furnace containing even 2% of hydrogen will cause only a very small hydrogen absorption which is not likely to exceed 0·5 c.c./100 g. It would appear, therefore, that the hydrogen content of industrial reheating-furnace atmospheres can be disregarded as an influence on the final hydrogen content of commercial steel.

ACKNOWLEDGMENTS

The author wishes to record his thanks to the Directors of The Clyde Alloy Steel Co., Ltd., for permission to publish this work, and in particular to Dr. R. Hunter for his help and encouragement in the course of its preparation.

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Correspondence on—
THE FIRST REPORT OF THE CONVERTER REFRACTORIES SUB-COMMITTEE*

REPLY TO DISCUSSION

The Chairman of the Converter Refractories Sub-Committee (Dr. R. J. Sarjant), on behalf of the Sub-Committee, wrote: The Sub-Committee is indebted to Mr. D. Howard Wood for his reference to the co-operative work involved in the Report. It was gratifying to the members to find such a ready response from the foundry industry as was the result of the circulation of that much criticized instrument, the questionnaire.

Dr. White's comments on the significance of the vitrified layer in monolithic linings are most apposite, and further investigations into the means of raising the viscosity of the vitrifying layer would undoubtedly widen the range of materials which could be regarded as suitable for the monolithic lining of converters.

The steep temperature gradient which has been found to exist in such linings under average working conditions is confirmed, as Dr. Rigby has so well explained, by study of the mineralogy of the linings. The cristobalite zone constituting the hotter portions of an open-hearth-furnace roof brick is a zone of high refractoriness, or expressed alternatively, a zone of high viscosity at high temperatures. It may be that the analogous zone of a converter lining has similar properties, but the distinction between monolithic linings and brick linings, which Dr. White mentioned, appears to be well confirmed though the reasons for the more extended vitreous layer and the lower oxide penetration of monolithic linings are imperfectly understood. Strict comparison is rendered difficult by the dissimilarity of the bricks and the monolithic materials studied so far in regard to properties such as composition, refractoriness, and previous thermal history.

In reply to Dr. Archibald's criticism relating to the data in Table I (*b*), regarding Plant *J*, 127 heats per run is a typographical error. The correct figure is 10–12. Table III, p. 240, deals with general trends, and as the vessel shape, refractories practice and consumption are so abnormal at Plant *N*, it was considered advisable to ignore these in assessing the general trend. The Sub-Committee feel justified in adhering to these figures, especially in view of new information supplied since the publication of the Report. Two firms whose refractories consumption in the 40–47-cwt. group is not included in Table II have now supplied details, and they fall in line with those given in column 4 of Table III.

The Sub-Committee were not as specific as Dr. Archibald asserts, in their remarks on clay content and permanent linear change. They only state that it appears to be an influencing factor, and they couple the nature of the clay with its amount. The suggestion of 90% of silica as a desirable minimum content was prompted by a desire to attempt a specific proposal. It should naturally be tied up with what the remaining 10% would consist of. It would almost certainly mean an increase in the alumina content to the eutectic point, with silica contents lower than 90%. The accumulated experience of the members of the Sub-Committee on this point was considered as well as the evidence presented in the Report. In their opinion the figure should be regarded as tentative until further evidence can be brought forward to justify its modification. Dr. Archibald's suggestion to use a crushing-strength test of preheated bricks from a dummy converter as a means of determination of the safe rate of heating, suffers from the limitation that it might give quite misleading indications as to the bearing of this factor on the ultimate life of the lining. The whole matter is bound up with many other factors such, for example, as the degree to which later fusion of the surface of the lining and slag penetration can impart strength to the lining.

Dr. Archibald appears to have overlooked one detail of Fig. 7 in attributing the higher mean temperature with sequential blowing to progressive wear of the lining. Reference to Fig. 7 will show that the peak temperatures are not rising all the time, but have reached practically constant values for heats 9, 10, and 11, the last three heats of the period of alternate blowing.

When sequential blowing of charges is commenced, it will be observed that the blow commences with the lining at a higher temperature (due to the shorter standing time) and, in the first heat of the series (Heat 12) a marked rise in the peak temperature is recorded. A further but smaller increase occurs in Heat 13 and actually a slight fall in Heat 14. In a foundry engaged solely on the mass production of relatively small quantities of hot steel at short time intervals, it is difficult to see how these observations can be explained in terms of progressive lining wear. Observations over an extended period (only portions are shown in the diagrams in the Report) show that whilst the effect of lining wear is discernible from day to day, it is almost completely masked by other effects over shorter periods.

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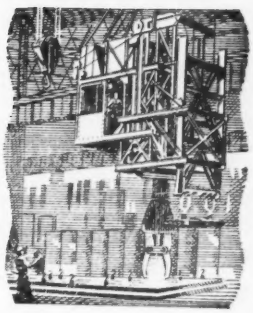
* For Discussion see *Journal of the Iron and Steel Institute*, 1947, vol. 157, Sept., pp. 99–114.

IRON AND STEELWORKS ENGINEERING



THE IRON AND STEEL ENGINEERS GROUP

THE CORNELLY LIMESTONE QUARRY OF THE STEEL COMPANY OF WALES, LTD.

A Description of Some Operating Difficulties *

THE development envisaged by the Steel Company of Wales, Ltd., at their Margam Works and the construction of the new strip mill at their Abbey Works called for a large building programme requiring rather more than a quarter of a million cubic yards of reinforced concrete. The aggregate for this and the demands for limestone for the steel production in the new plant made it necessary to mechanize the firm's limestone quarry at Cornelly.

The new mechanized plant was designed by the International Construction Company, Ltd., consulting engineers to Messrs. Guest Keen Baldwins Iron and Steel Co., Ltd., in conjunction with Messrs. Fraser and Chalmers, Ltd., who were responsible for the erection.

Work was begun by Sir Robert McAlpine and Sons, Ltd., the civil engineering contractors, on 1st October, 1945, and by Messrs. Fraser and Chalmers, Ltd., in May, 1946. Constructional work was completed during December, 1946, and the new plant was put into production on 1st January, 1947, the change-over from the old method of working to the new being done in two days. A full description of the plant was given in the *Proceedings of the South Wales Institute of Engineers* in April, 1947, extracts of which were published in the *Iron and Coal Trades Review*, of May 23rd, 1947.

The layout of the new plant presented difficulties owing to the restricted space in the quarry and the very broken nature of the ground because of old workings, and the fact that the old railway access to the quarry had to be maintained during the construction added to the difficulties. A general picture of the plant is given in Fig. 1.

The plant was originally designed for a total output of 7000 tons per week with single-shift working. The demand for limestone, however, has exceeded the estimates and it has been necessary to change the plant on to two-shift operation. Output has gradually risen, the present record output being 9700 tons in a week. It is hoped, when the length of the face is greater, to increase the weekly output to 11,000 tons.

When the building of the new works is completed, approximately 6000 tons of limestone per week will be needed in the manufacture of iron and steel; the remainder of the quarry output will then be available for sale.

FIG. 1—General view of quarry

* Received 31st October, 1947.

Removal of Overburden

The overburden covering the area now being worked consists of clayey soil from 3 to 7 ft. deep, with old clay pockets penetrating into the stone. This overburden was at first removed by a 9-cu. ft. Ruston Bucyrus excavator with drag-line equipment, using a $\frac{5}{8}$-cu. yd. bucket loading into 1-cu. yd. 2-ft. gauge tipping skips, 20 skips being linked together. These skips ran on narrow-gauge railway track to a disposal dump some 600 yards away.

It was found that owing to the time taken for the locomotive to haul the skips to the disposal dump, empty them, and return them to the operating point the excavator was in fact only working at about one-third of its capacity. It was therefore decided to take full advantage of the fine weather and to strip during the summer sufficient top soil and broken top stone (to which the clay adheres) to tide over the winter months, when conditions at the quarry top are bad in any case. A *D*.8 bulldozer with a 12-yd. scraper has been used to remove the top soil, followed by a three-tine rooter to break up the top stone, clearance of which down to the top beds is effected with a further use of the scraper. This method has proved successful. Furthermore, by using a *D*.8 bulldozer it was possible to clear beforehand a suitable dumping area within economical working distance.

Drilling

Drilling is carried out by three Ruston Bucyrus 27*T* blast-hole drilling machines, specified as capable of drilling $6\frac{5}{8}$-in. dia. holes to a maximum depth of 200 ft., at a rate of 3 ft./hr. in limestone.

Under good conditions a maximum rate of 5 ft. 9 in./hr. has been reached. Allowing for changing bits, moving up, adjustments, etc., an average of 3 ft./hr. has been maintained. As far as the drilling machines themselves are concerned no mechanical or electrical troubles have been experienced, but some difficulty has been encountered with drilling stems breaking off at the boxes.

Charging of Holes

The old method of working the face down in benches had resulted in considerable " toes " being left which had to be tackled in order to obtain the vertical face essential for the new method of working. In the initial stages holes were spaced 10 ft. apart and as near as possible to the edge of the face. The height of the face being about 80 ft., the actual holes were drilled to a depth of 82 ft.

Open-cast gelignite was first tried but was not powerful enough to kick out the toes; the blast travelled up the hole following the clay backs, prevalent in this face, causing a bad break back and shattering the ground badly. Trouble with drill stems breaking at the boxes was chiefly caused by these conditions, since the drills kept running off at a depth of about 40 ft. into cracks and clay pockets, which put abnormal strain on the boxes. Various methods were tried for straightening the holes when this occurred. Broken fire-bars inserted in the hole were, on occasions, successful, but it is felt that now that ground conditions have improved no further trouble should arise.

Blasting gelatine dynamite was substituted as a base charge and gelignite as a column charge. Shattering of the toe is still not satisfactory but there is very much better fragmentation of the bulk of the stone. No failures have occurred using Cordtex instantaneous fuse as a detonating agent.

Secondary Blasting

Air is supplied to the pneumatic drills for pop-shooting by an Alley and McLellan Sentinel Compound Interlocking Air Compressor No. 5, of capacity 400 cu. ft./min. at 100 lb. sq. in. pressure, the 2-in. dia. air pipes being carried in the ground, 1 ft. 6 in. below ground level. It has been found necessary to insert air filters in the air line to prevent freezing-up under severe conditions.

The 100 Ruston Bucyrus $3\frac{1}{2}$-cu. yd. Excavator

The Ruston Bucyrus $3\frac{1}{2}$-cu. yd. excavator has been found capable of a very high hourly output, provided that it has a large pile of stone into which to dig. No electrical or mechanical faults have been experienced with the exception of the fracture of dipper sticks, which had to be welded.

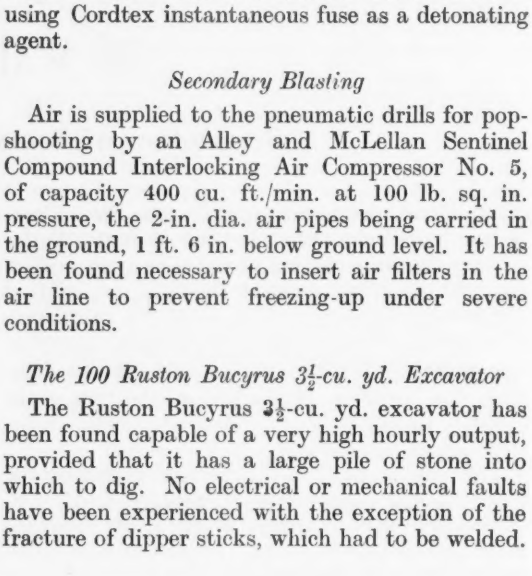


FIG. 2—Quarry dump truck

It has, however, been found necessary to replace the $\frac{5}{8}$-in. mild-steel plate round the cab by $\frac{3}{4}$-in. armour plating, since on two occasions falls of stone have penetrated the cab and damaged the excavator equipment.

The 12-ton Foden Dump Trucks

The 12-ton Foden dump trucks (*see* Fig. 2) are operating most satisfactorily. No trouble has been experienced with the engines or chassis, but the tipping gear, which is hydraulically operated by a four-way pump driven off the gear-box, is too slow: the makers are redesigning this in new models to enable the tipping cycle to be completed in less than one minute.

The shape of the body, which was designed by Guest Keen Baldwins with parallel sides and a tail sloped at 161°, does not allow the load to empty itself until a very steep angle of tip is reached; the load then slides off all at once over the tail, causing the front of the truck to lift. Some improvement has been obtained by altering the slope of the tail to 170°. In the later designs the wagon sides will be splayed, and the open end will be wider than the front end to avoid any danger of wedging action during tipping.

It has also been necessary to increase the length of the protection plates over the driver's cab to avoid damage to the cab roof, radiator top, and headlamps when loading: later designs will include additional protection for the diesel fuel tanks.

The Ross Drop-Bar Feeder

The Ross drop-bar feeder consists of an apron feed, driven by sprockets through a worm reduction gear. It is supported by a mild-steel frame, with four shock-coil springs transferring the shock to two transverse members, connected at both ends to 16 × 8-in. rolled steel joists. These longitudinal members in turn rest on two 24 × $7\frac{1}{2}$-in. rolled steel joists. The design of the hopper is such that the dump trucks tip directly on to the apron. During inspection, the transverse members (consisting of two 8 × $3\frac{1}{2}$-in. rolled steel channels, back-to-back with $\frac{1}{2}$-in. mild-steel spacer plates at either end) were found to be deflected. Mild-steel plate, $\frac{3}{4}$-in. thick, was welded underneath to prevent further distortion. Shortly after this the bolts connecting the transverse members to the 16 × 8-in. rolled steel joist sheared. The bolts were replaced, but continued to shear off, deflection increasing until the chassis collapsed. Fillets welded in to stiffen this were not successful and 10 × $3\frac{1}{2}$-in. rolled steel channels were substituted for the original 8 × $3\frac{1}{2}$-in. ones. In addition a full length of $\frac{1}{2}$-in. web plate was inserted, giving a full web thickness of $1\frac{1}{2}$ in. The channel connections to the 18 × 8-in. longitudinal joists have been made very robust. The 16 × 8-in. longitudinal rolled steel joist seated on top of the 24 × $7\frac{1}{2}$-in. rolled steel joist was reinforced in the web by $\frac{1}{2}$-in. stiffeners at points where the transverse channels connected into it. Up to the present this has been successful.

The Primary-Crusher Main-Feed Hopper

In the early stages, large pieces of stone frequently became wedged against the front plate of the primary-crusher main-feed hopper, resulting in both sides of the back end of the hopper splitting open for a distance of 12 ft. The size of the front opening of the hopper was increased, both sides were welded up, and four butt straps fixed at 4 in. apart on either side. No jams have occurred at this point since.

The Two-Roll Grizzly

The drop from the drop-bar feeder to the two-roll grizzly is about 16 ft. Large stones were liable to fall off the feeder on to the front roll, causing fractures of the roll segments, bending of the roll shaft, and even breaking the roll-supporting pedestal and bearing housing. Hardwood packers were inserted between the segments of the rolls, but in spite of this another roll segment was subsequently broken. A very heavy chain curtain was then suspended immediately in front of the drop-bar feeder. This has resulted in a very great improvement in the feed to the grizzly, since the heavy stones are held back and drop on to the back roll. It has also improved the screening efficiency of the grizzly.

The Primary Crusher and Chute

The primary crusher is yielding an output far in excess of the maker's stated tonnage. Some initial difficulty was encountered through the bearings running hot. This has since been overcome by changing the lubricant. Much output was lost owing to jams in the primary chute caused by the stone breaking into long slivers which lay along the chute where it tapered down to feed on to the roller feeder. This has been overcome by widening the chute and reducing the height of the straight portion above the roll feed.

Intermediate Screen and Picking Section

In the original scheme (*see* Fig. 3), the finest material screened out on the drop-bar feeder and two-roll grizzly was fed on to a 24-in. conveyor (No. 3), running at 200 ft./min. and with a capacity of 130 tons/hr., up to the intermediate screen house, where the belt delivered its load on to a single-deck Gyrex screen, with a 2-in. square mesh mat. This screen separated out the 2-in. material, which was fed on to No. 4 conveyor, running at 80 ft./min. and designed as a picking section for the removal of clay. The + 2-in.

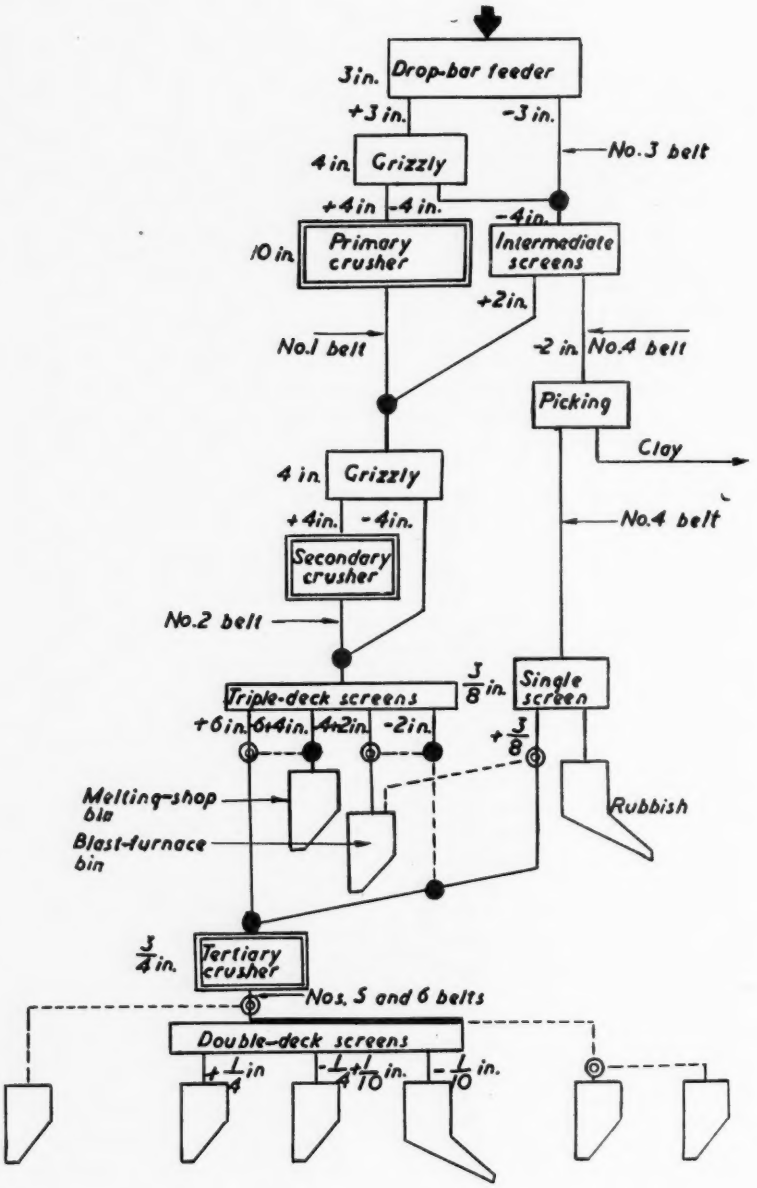


FIG. 3—Original flow sheet

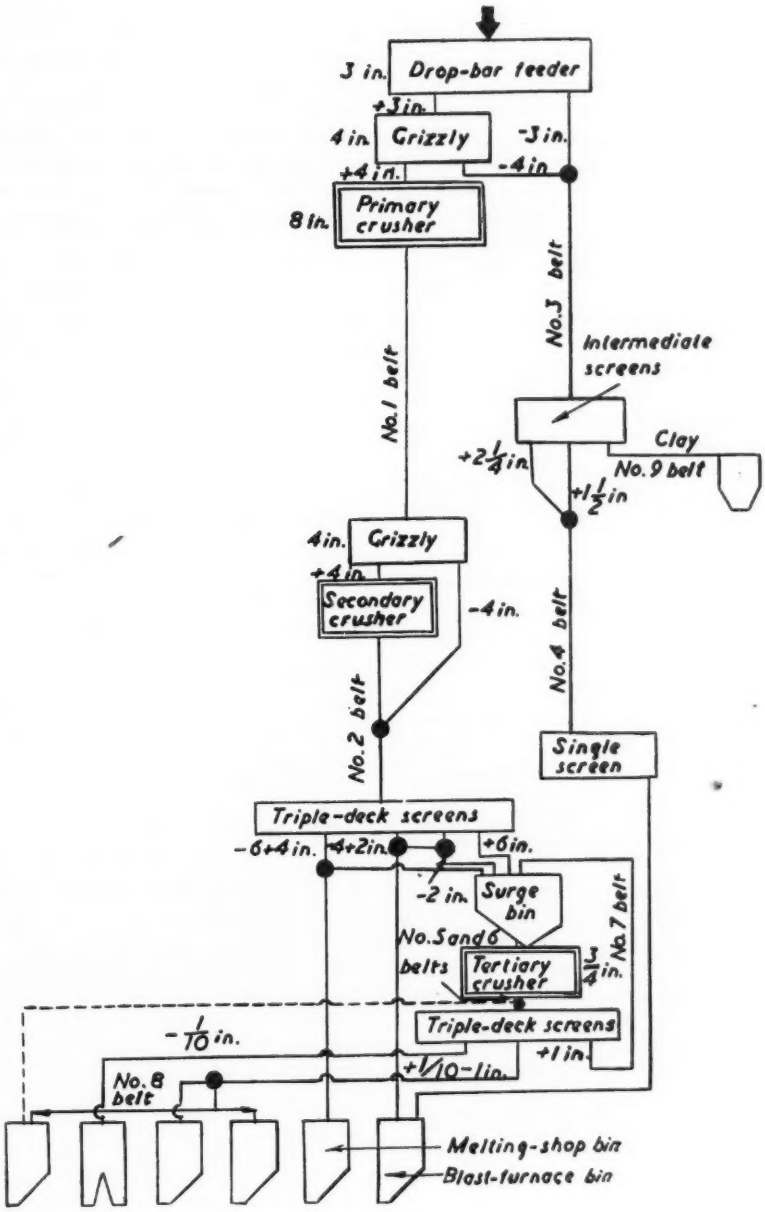


FIG. 4—New flow sheet

XUM

material was fed by a chute on to No. 1 conveyor, which was carrying the primary-crusher product. The amount of clay was far greater than anticipated and picking was found to be out of the question. Belt No. 3 was at times overloaded, and so was speeded up to 300 ft./min.; the speed of No. 4 belt was increased from 80 to 120 ft./min.

Modifications are now in hand to dispose of the dirt at the intermediate house (*see* Fig. 4). They are as follows: Because of the very large amount of dirt from the face passing up No. 3 belt, the screening at the intermediate house has been split by the insertion of a double-deck screen in place of the original single-deck screen. The "overs" from both decks are now taken up No. 4 conveyor direct to the blast-furnace bunker; the "throughs" are taken by a new 24-in. dirt belt to an open-fronted bunker, from which the dirt is fed by two 36-in. heavy-type Sherwin feeders into lorries for dumping. Consequently the stone line through the second and third crushers does not get any feed except broken lump stone from the primary, and it is hoped that this will give a very much cleaner aggregate.

Conveyors and Secondary Crusher

The conveyors, which run at 300 ft./min., are satisfactory. The smallest tears in the belt are repaired immediately, by the use of portable vulcanizing units. The secondary crusher, which is a Pennsylvania single-roll 30 × 70 crusher, has given no trouble, except for initial lubricating difficulties.

The Tertiary Crusher

The tertiary crusher is a 4½-ft. Symons cone, and has been free from trouble. The product, however, contains rather a high percentage of flats; it is hoped to improve this by choke feeding, as follows: The original feed to the tertiary crusher was through a long closed chute taking either — 2-in., — 2-in. and + 6-in., or — 2-in. and + 4-in. material. It was found that — 2-in. stone was insufficient for choke-feeding the tertiary, whereas if either of the other products was added the chutes became choked. A surge-bunker has therefore been installed in place of the long chute, with a Sherwin feeder discharging directly into the tertiary. The feeder is automatically operated by the amount of stone in the surge-bin. It is hoped that this arrangement will enable any product to be passed through the tertiary at will, thus maintaining its output at a maximum.

The Ropeway

The ropeway was built for an output of 150 tons/hr. If the buckets are filled exactly to the required level, it will actually carry this load. Difficulty has been found with buckets not gripping the haulage rope properly, causing them to tip on the run down and sometimes to come off the haulage rope altogether. On starting a serious surge was set up, undoubtedly increasing any minor adjustment troubles. The guide rail at the dispatch end was adjusted and a slip-ring motor substituted for the original high-torque squirrel-cage motor, giving much smoother starting. The ropeway has run for a month since the change-over without a recurrence of trouble.

The Bins

The chutes for the bins required slight modification; the chutes from the 6-in. bunker feeding the ropeway in particular gave trouble by the stone arching above the chutes. A heavy-type Sherwin vibrator helped to cure this: the trouble has not, however, recurred since the secondary crusher was set down.

Dirt Handling

The present system of handling dirt is through a closed bunker, with a quite severe taper to the chute. This has been a source of serious trouble in wet weather as the material, when wet, sticks even to the vertical bunker or chute sides; it very quickly builds up and cements itself into a solid mass, completely closing up the bunker or chute. This clogged material can then only be moved by a man with a pick-axe, necessitating the closing of the plant for two or three hours at a time. During dry weather there is no trouble at all.

The open-fronted bunker now being installed will, it is hoped, much improve the position, as the sides have been designed with a very small taper and delivery is by a full-width Sherwin feeder.

It is necessary to handle this wet and dirty material through chutes. In the winter of 1946–47, heavy vibrators were put on the chutes concerned with some success; it was, however, found necessary at times to station men at various points on the chutes to ensure that no build-up started.

Dust

Dust is a seasonal problem. During wet weather it is not serious, but during a long period of dry weather some adequate means of suppressing it is necessary. Various methods of dust collection are being installed, but it is not yet known how successful they will be. A comprehensive system of inspection and daily task maintenance is being successfully operated on the whole plant, but the maintenance difficulties would be greatly relieved if the problem of dust collection were solved.

N E W S

ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE

Autumn General Meeting, 1947

The Autumn General Meeting of the Iron and Steel Institute was held on Wednesday and Thursday, 12th and 13th November, 1947, under the chairmanship of the President, Dr. C. H. Desch, F.R.S., at the Offices of the Institute, 4, Grosvenor Gardens, London, S.W.1.

President-Elect for 1948

It was announced that the Council has nominated as President, to take office in May, 1948, Sir Andrew McCance, D.Sc., F.R.S., Deputy Chairman and Joint Managing Director of Messrs. Colvilles, Ltd.

Forthcoming Meetings

The Secretary, after giving details of changes on the Council, announced the following dates of future meetings :

Annual General Meeting, 1948 : Wednesday and Thursday, 5th and 6th May, 1948.

Blast-Furnace Symposium : Wednesday and Thursday, 26th and 27th May, 1948.

Autumn General Meeting, 1948 : Wednesday and Thursday, 10th and 11th November, 1948.

In addition a special summer meeting for members, and particularly for Continental members, during the end of the first and during the second week of July, 1948, is under consideration. Details will be announced later.

Remarks by the President

In his opening remarks Dr. C. H. Desch, F.R.S., said that it was anticipated that this would be a relatively small meeting, owing to the prevailing conditions, but a small meeting had the advantage that it could be held in the Institute's own building, and there was no reason why the discussions should not be excellent. On behalf of the Council, Dr. Desch welcomed all the members and visitors present, especially those from oversea.

Dr. Desch thought that the Institute could look back with some satisfaction on the work of the year now drawing to a close. In addition to the Annual Meeting in May, a very successful special Summer Meeting had been held in Switzerland in July ; the Symposium on Powder Metallurgy, held last June, had been the first meeting of its kind to be held in this country. A joint meeting had been held with the British Iron and Steel Research Association and the Institute of British Foundrymen in March, and a further meeting with the British Iron and Steel Research Association in June. The Iron and Steel Engineers Group had been very active ; three meetings would have been held by the end of this year and four meetings had been planned for next year.

A full report of the Autumn Meeting will be published in the March, 1948, issue of the Journal.

Interchange of Abstracts with Jernkontoret

An arrangement, to take effect as from 1st January, 1948, has been made with the Jernkontoret, Sweden, whereby abstracts published in the *Journal* will be exchanged for abstracts of Scandinavian literature of interest to members of the Institute. These abstracts will be published from time to time in the Abstract Section of the *Journal*. This will greatly increase the range of technical literature available to members.

EDUCATION

National Certificates in Metallurgy

The Joint Committee for National Certificates has issued its Report for the year 1946–1947, and states that the Scheme for the award of National Certificates in Metallurgy continues to make good progress.

Final examinations were held this year at the following Colleges :

Ordinary Certificate—Battersea Polytechnic, London; Birmingham Central Technical College ; Chesterfield Technical College ; Coventry Technical College ; Merchant Venturers' Technical College, Bristol ; Rotherham Technical College ; Scunthorpe Technical School ; The Chance Technical College, Smethwick ; Swansea Technical College ; The County Technical College, Wednesbury ; and Wolverhampton and Staffordshire Technical College.

Higher Certificate—The County Technical College, Wednesbury ; and Wolverhampton and Staffordshire Technical College.

Distinctions were awarded to thirteen candidates who have shown an exceptional grasp of their subjects, indicating a high degree of training and knowledge in the particular subject in which the distinction was gained. Prizes, taken in books and distributed by the Colleges concerned, were awarded to seven successful candidates.

Staff Biography No. 12

Dr. NORMAN STUART, Education Officer, was born in London in 1907 and educated at Heath Mount School, Hampstead, at Dover College, and at the Imperial College of Science and Technology, where he took an honours chemistry course and graduated in

1928. Subsequent research work in collaboration with Professor Sir George Thomson formed the subject of a thesis for the degree of Ph.D.

From 1932 to 1941 Dr. Stuart was engaged in teaching and during these years held appointments at the Royal College of Science, Shoreditch Technical Institute, Stowe School, Rugby School, and the L.C.C. Training College for Teachers. In 1941 he went to Cambridge to assist in experimental work for the Ministry of Supply. Later in that year, at the invitation of Dr. U. R. Evans, he joined the team of workers at the Corrosion Laboratory and carried out an experimental investigation upon the influence of zinc as a cathode protector against the corrosion-fatigue of mild steel. In 1942 he returned to the Ministry of Supply and was appointed an experimental officer responsible for foreign and imperial liaison on scientific research and development.

In the following year Dr. Stuart was seconded to the Royal College of Science to assist Dr. A. S. C. Lawrence in various experimental problems for the Ministry of Supply and the Admiralty. When this work was discontinued in 1944, he took up an appointment with the Fire Research Division of the Ministry of Home Security, and became a member of the Incendiary Bombs Test Panel.

Dr. Stuart took up his present appointment as Education Officer to the Iron and Steel Institute in March, 1946, since which time he has been concerned mainly with the various activities of the Joint Committee on Metallurgical Education and with the Joint Committee on National Certificates in Metallurgy.

Apart from his metallurgical and general scientific interests, Dr. Stuart has always been a keen and active amateur musician and is also interested in colour photography. While Honorary Secretary to the Royal Institute of Chemistry, Students' Association, he founded and co-edited the first three volumes of the journal *Monochrome*.

NEWS OF MEMBERS

▶ Mr. S. M. BARKER has left Messrs. Richard Thomas and Baldwins, Ltd., to take up an appointment in the Metallurgy Division of the National Physical Laboratory, Teddington.

▶ Dr. J. H. CHESTERS, of the United Steel Companies, Ltd., has moved from Stocksbridge, near Sheffield, to Swinden House, Moorgate, Rotherham.

▶ Mr. D. K. COUTTS is now Chief Metallurgist at the Indian Iron and Steel Co., Ltd., Kueti Works, Kueti, E.I. Railway, Bengal, India.

▶ Capt. H. LEIGHTON DAVIES has resigned his directorship of Baldwins (Holdings), Ltd., and has been succeeded by Col. J. M. BEVAN. Mr. ANDREW AGNEW has become Vice-Chairman.

▶ Mr. E. J. EVANS has left Messrs. Stewarts and Lloyds, Ltd., to take up an appointment as Works Metallurgist and Chemist at the Welsh Tinplate and Metal Stamping Co., Ltd., Llanelly.

▶ Mr. F. HARGREAVES, now at the Armament Research Department, Ministry of Supply, is leaving shortly to become Metallurgist to Philips Blackburn Works, Ltd., Blackburn.

▶ Mr. E. T. JUDGE, of Messrs. Dorman, Long and Co., Ltd., has changed his address to G.P.O., Box No. 11, Royal Exchange, Middlesbrough.

▶ Mr. KONRAD KORNFELD has taken up a teaching appointment with the Polish Merchant Navy College, Landywood, near Walsall, Staffs.

▶ Dr. A. MCCANCE, F.R.S., President-Elect of the Iron and Steel Institute, has resigned from the Advisory Council of Scientific and Industrial Research.

▶ Mr. J. MCMULLEN has left C.A.V., Ltd., and is now Chief Metallurgist with the Electric Resistance Co., Ltd., Weybridge.

▶ Mr. H. H. MARDON, Superintendent of the Plant Engineering Division, B.I.S.R.A., has been transferred to the class of Member of the Institution of Mechanical Engineers.

▶ Mr. A. PREECE, President of the Leeds Metallurgical Society, which is affiliated to the Iron and Steel Institute, has accepted an invitation to become an Honorary Member of Council of the Institute during his period of office.

▶ Mr. J. A. ROBERTS has graduated from Swansea University with honours in metallurgy and is now at the aluminium works of Messrs. Richard Thomas and Baldwins, Ltd.

▶ Mr. A. D. ROBERTSON has left the Workington Iron and Steel Co., Ltd., to join the United Steel Companies, Ltd., Research and Development Department, Swindon House, Moorgate, Rotherham.

▶ Mr. GERALD STEEL has been appointed Joint Managing Director, with the Chairman, Sir Walter Benton Jones, Bt., of the United Steel Companies, Ltd.

▶ Mr. ASHLEY S. WARD, Chairman of Messrs. Thos. W. Ward, Ltd., and other associated companies, has relinquished the chairmanship of Lowmoor Best Yorkshire Iron, Ltd., and Mr. GEORGE WOOD, Deputy Chairman of Messrs. Thos. W. Ward, Ltd., has been appointed in his place. Mr. C. L. FRY and Mr. R. STUBBS become Joint Managing Directors, and Mr. W. W. HICKMAN has been appointed a Director.

▶ Mr. G. H. H. WILLIAMS has left the English Electric Co., Ltd., Stafford, and has returned to University College, Swansea.

Elections

▶ Professor PIERRE CHEVENARD has been elected an Honorary Member of the Institute of Metals.

▶ Dr. C. H. DESCH, F.R.S., President of the Iron and Steel Institute, has been elected a Fellow of the Institute of Metals.

▶ Sir WILLIAM GRIFFITHS, Chairman and Managing Director of the Mond Nickel Co., Ltd., has been appointed a Member of the Advisory Council for Scientific and Industrial Research.

▶ Mr. W. M. LARKE has been elected President of the Staffordshire Iron and Steel Institute.

Awards

▶ Dr. C. A. EDWARDS, F.R.S., has been awarded the honorary degree of LL.D. by the University of Wales.

▶ Mr. CHARLES R. HOOK, President of the American Rolling Mill Company, Middleton, Ohio, has been elected to receive the Medal for the Advancement of Research, for 1947, of the American Society for Metals.

▸ Mr. A. B. KINZEL, Vice-President of the Electro-Metallurgical Company and the Union Carbide and Carbon Research Laboratories, Inc., who delivered the 1947 Campbell Memorial Lecture of the American Society for Metals, has been awarded the Samuel Wylie Miller Memorial Medal.

▸ Mr. D. E. J. TALBOT has been awarded an honours degree in metallurgy of the University of Wales.

Memoranda

▸ British Coated Sheets, Ltd., are transferring their works from the Orb Ironworks, Newport, Mon., to the Mersey Ironworks, Ellesmere Port, Wirrall, Cheshire.

▸ The Research Laboratories of Messrs. Joseph Sankey and Sons, Ltd., Manor Works, Ettingshall, Wolverhampton, have now become the Central G.K.N. Research Laboratories, Manor Works, Ettingshall, Wolverhampton, Staffs.

▸ Messrs. Thos. W. Ward, Ltd., Sheffield, have formed a new Belgian organization; registered as Thos. W. Ward (Belgium) S.A., the offices have been opened at 8, Longue Rue des Claires, Antwerp.

Obituary

The Council regret to record the death of:

Mr. C. W. KAYSER, J.P., Chairman of Messrs. Kayser, Ellison and Co., Ltd., until 1946, on 23rd November, 1947, aged 77.

Mr. J. S. G. PRIMROSE, Chief Metallurgist, The Rover Co., Ltd., Coventry, on 2nd October, 1947.

Members' Addresses

The present addresses of the following members are unknown to the Institute's Records Department. Will these members, or others who can supply relevant information, please get in touch with the Secretary:

BURBAGE, Harold William Raymond
BEACH, Norman
CUSH, J. J., Jun.
DAVISON, Vivian Westgarth
FAIRBAIRN, Joyce
FERGUSON, William
FITZGIBBON, John J.
GRAY, Edwin Reed
GRAY, Kenneth
GREGORY, Clifford H.
GUEVARA, Ricardo
HARRISON, John William
HARRISON, Leslie
IHRE, Nils E.
LARN, John David
LE TALL, Sydney Harold
LEWIS, Michael Frederick Paul
MACKAY, Murray
MCLAY, Gavin Smellie
MACHALSKI, Joseph
MALMBERG, Rudolph E.
MARINKOV, Cedomir B.
MILLWARD, Edward Bertram
MUIR, Archibald Robert
NICHOLLS, William
OLSZAK, Feliks
PERKIN, Thomas Scott
PARAKH, Ardeshir Kaikhusru
PODGORSKI, Stefan
READ-PRYCE, Alfred William
ROBERTS, Eric William
ROGERS, Thomas Charles
ROLLIN, Calvert Noble
SAIN-MILETITCH, Branko
SHARP, Richard Bruce
SIMPSON, Francis Frederick Le Souef
SMITH, Ralph Douglas
TANN, Jiff P.
THOMAS, Ivor
WALTON, Frank
WARD, The Hon. Edward Frederick
WILLIAMS, Isaac
WILLOCK, Edward Garley
WRIGHT, Harold Vernon

CONTRIBUTORS TO THE JOURNAL

Georges Robert Delbart was born in 1899 at Anzin (Nord), France. A Doctor of Science and Engineer-Chemist of Lille University, from 1922 to 1927 he was Head of the Laboratory of the Escaut et Anzin, and from 1927 to 1946 was with the Anciens Etablissments Cail, at Denain (Nord), where he was successively Chief of the Laboratories, Manager of the Steelworks, and Chief Engineer of the Metallurgical Department.

During his 25 years of works experience Monsieur Delbart has striven to bridge the gap between the laboratory and the works, particularly at the Etablissments Cail, where he installed armament shops for the manufacture of large forgings, laminated and cast armour-plating, and thin sheets, and a foundry for high-quality and special cast irons.

Monsieur Delbart was in April 1946 appointed Director of the French Institut de Recherches de la Siderurgie; his duties are to set up the research laboratories for this Institute, of which the construction has been started at St. Germain-en-Laye.

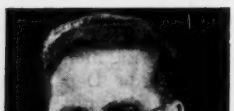
Monsieur Delbart has had published, in French and other journals, about eighty notes, memoirs, and papers, in which he has recorded the results of his laboratory investigations and of his experience in steelworks.

During the war Monsieur Delbart, a specialist in cast armour for tanks, was attached to the Ministry of Armaments as a scientific adviser, and in this rôle took part in a mission to the Vickers steelworks. He has been a member of the Iron and Steel Institute for more than ten years, and has many friends in England.

Rubin Potaszkin—Chief of Technical Metallurgical Departments at the Société Française de Constructions Mécaniques, Anciens Etablissements Cail, at Denain (Nord). Monsieur Potaszkin, who is of Russian origin and a naturalized Frenchman, was born in Vilna in 1905. He was educated at Eupatoria (Crimea) at Strelitz Technical School, Mecklenburg, and at Toulouse University. Employed before the war as an engineer in the turbines and centrifugal compressor department of the Société Rateau, he served with the French Army from 1939 to 1945 and was awarded

G. R. Delbart

R. Potaszkin



D. C. Muir

J. Cameron

M. P. Newby

D. McLean

the Croix de Guerre. He has had a large number of technical papers published, most of which were written in collaboration with Monsieur G. Delbart.

Dr. P. Herasymenko—Born in 1900 in Odessa, Ukraine, Dr. Herasymenko emigrated to Czechoslovakia in 1921 and studied chemistry and physics at the Czech Charles' University, Prague, where he graduated in 1925. Later he became a research student in physical chemistry at King's College, London University. In 1929 he was appointed lecturer in physical chemistry at the Ukrainian Free University in Prague, and in 1930 became the research physical chemist and metallurgist at the Research Institute of the Skoda Works in Pilsen, still continuing to lecture at the Ukrainian Free University.

During the German occupation Dr. Herasymenko was imprisoned by the Gestapo for nearly a year, and is at present living as a refugee in Munich.

He joined the Iron and Steel Institute in 1938, and has written a number of papers on electrochemistry, physical chemistry of steelmaking, and metallography.

J. Cameron, Ph.D., B.Sc., A.R.I.C.—In charge of research at the Clyde Alloy Steel Co., Ltd. Dr. Cameron was born at Sorn, Ayrshire, and was educated at Kilmarnock Academy, and at Glasgow University, where he graduated in 1936 with a first-class honours degree in physical chemistry. He was awarded a D.S.I.R. research scholarship and later a Carnegie Trust research scholarship which enabled him to pursue post-graduate research for which he obtained the Ph.D. degree in 1939. In that year he joined the staff of the Metal Gas Co., Ltd., Motherwell, where he was concerned with the gaseous chromizing process. He joined the research department of the Clyde Alloy Steel Co., Ltd., in 1940, and was appointed to his present position early in 1947.

Dr. Cameron has written papers on photo-chemistry and spectroscopy. He serves on a number of sub-committees of the British Iron and Steel Research Association.

D. C. Muir—Ceramic Engineer and Refractories Technologist at the Consett Iron Co., Ltd. Mr. Muir was educated at Monkseaton Grammar School and at Rutherford College, Newcastle, and trained as chemist and gas engineer under the late Dr. Weyman at the various works of the Newcastle and Gateshead Gas Company.

In 1929 Mr. Muir joined the Consett Iron Co., Ltd., as Chief Chemist at their Derwenthaugh Coke Works, and was subsequently Technical Representative of the Company for three years, during which time he visited America to investigate the coke-handling facilities at the eastern ports. He was transferred to the Refractories Department in 1936 to co-ordinate and develop the research work in the steel and brick works. During the war he also acted as Manager of the ganister quarries and of the company's silica, basic, and firebrick works.

Mr. Muir is a member of various technical societies and serves on the Steelworks Refractories Joint Committee and several sub-committees of that body.

M. P. Newby, B.Sc., A.Inst.P.—In charge of the Aerodynamics Section of the Physics Department at B.I.S.R.A. Mr. Newby was born at Erith, Kent, in 1917 and received his education at Deacon's School, Peterborough, and from 1937 to 1940 at University College, Nottingham, where he obtained an external degree of B.Sc. in physics, of London University.

During the war Mr. Newby served in the R.A.F.V.R. and was commissioned into the Meteorological Section in April 1943. He left the R.A.F.V.R. in June 1946, with the rank of Flight Lieutenant, and took up his position with B.I.S.R.A. in October of that year.

D. McLean was born in 1915. He was educated at Manchester Grammar School and at Manchester University where he studied metallurgy under Professor Thompson, graduating in 1936. Mr. McLean spent two years with the Imperial Smelting Corporation, working on problems connected with zinc. Later he joined the Armament Research Department and became interested in the field of physics of metals. He transferred to the National Physical Laboratory in this year.

IRON AND STEEL ENGINEERS GROUP

Future Meetings

London

The **Sixth Meeting** of the Iron and Steel Engineers Group will be held at 4, Grosvenor Gardens, London, S.W.1, on Wednesday, 18th February, 1948.

The morning session will be devoted to the presentation and discussion of a paper on "*Engineering Aspects of the Distribution and Utilization of Gaseous Fuels in Iron and Steel Works*," by J. B. R. Brooke and J. S. Bryan (J. Lysaght, Ltd.). During the afternoon session a paper on "*Soaking Pits*," by the Society of Furnace Builders, will be presented and discussed.

Sheffield

A Joint Meeting of the Iron and Steel Engineers Group with the Sheffield Society of Engineers and Metallurgists will be held at the Royal Victoria Station Hotel, Sheffield, on Monday, 19th January, 1948, at 6.15 P.M. The President of the Sheffield Society of Engineers and Metallurgists will be in the Chair.

The meeting will be devoted to the presentation and discussion of the paper entitled "*Mechanical and Electrical Features of Primary Hot - Rolling - Mill Auxiliaries*" by W. W. Franklin, of Davy and United Engineering Co., Ltd., and P. F. Grove, of Messrs. John Miles and Partners (London), Ltd., which was published in the November issue of the *Journal*.

All members of the Group who wish to attend this meeting will be welcome.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Oxygen-Enriched Air in Side-Blown Converters

The Association has issued a paper entitled "*The Application of Oxygen-Enriched Air to Side-Blown Converter Practice*," by J. L. Harrison and A. Hartley. The authors demonstrate the great possibilities of this process and show how large economies are obtained by the addition of quite small quantities of oxygen to the air blast.

In view of the topical nature of this report, early attention is being drawn to the paper, which is available at present to members of the Research Association. Every endeavour will be made to publish the paper in the BISRA section of an early issue of the *Journal*.

NEWS OF SCIENCE AND INDUSTRY

Indian Institute of Metals

The Indian Institute of Metals will be inaugurated on Monday, 29th December, 1947, at the Royal Asiatic Society's Hall, Park Street, Calcutta. The Honourable Dr. Syama Prasad Mookerjee, Member for Industry and Supply, Government of India, has kindly consented to inaugurate the Institute, and Sir Jehanjir Ghandy, Director of Tata Industries, Ltd., will deliver the Presidential Address.

The Iron and Steel Institute will be officially represented by Dr. D. P. Antia, who as honorary secretary took a large share in organizing the Indian Institute of Metals. Members of the Institute are cordially invited to be present at the meeting.

National Foundry College

The National Foundry College has now been established, and its Board of Governors, under the Chairmanship of Mr. Randall G. Hosking, has been nominated by the Ministry of Education. The College will be housed in, and work in close conjunction with, the Wolverhampton and Staffordshire Technical College.

Facilities will be provided for foundry education and research, in the form of full-time courses, vacation courses, refresher courses, and part-time day and evening courses. A Diploma, endorsed by the Ministry of Education, will be awarded to successful students.

The Ministry of Education, on the recommendation of the Board of Governors, has appointed Mr. J. Bamford, B.Sc., to be Head of the College.

Details in the form of a list of the Board of Governors and the bodies represented, prospectus and curriculum, and a form of application for admission, are now available and may be obtained on request.

Sheet and Strip Rolling

A course of six lectures will be given in the School of Industrial Metallurgy, University of Birmingham, on "*An Introduction to the Theory of Sheet and Strip Rolling, and Its Application to Industrial Problems*," by Dr. A. R. E. Singer. These lectures, which will take place on successive Saturday mornings, commencing on 31st January, 1948, at 10.30 A.M., are planned to meet the requirements of senior executive officers, metallurgists, engineers, and technicians, in the metal-producing and metal-consuming industries. The syllabus will cover theories applicable to the rolling of sheet and strip, and the factors governing the development and choice of rolling-mill plant.

DIARY

1948

5th January

CLEVELAND INSTITUTION OF ENGINEERS—Lecture on "*Desiliconization*," by E. F. Brown—Cleveland Scientific and Technical Institute, Corporation Road, Middlesbrough, 6.30 P.M.

8th January

LEEDS METALLURGICAL SOCIETY—Symposium on Casting—Chemistry Department, Leeds University, 7.0 P.M.

9th January

MANCHESTER ASSOCIATION OF ENGINEERS—"*Researches now in Progress in the Cavendish Laboratory*," by Sir Lawrence Bragg, O.B.E., M.C., D.Sc., F.R.S.—College of Technology, Manchester, 6.45 P.M.

10th January

SWANSEA AND DISTRICT METALLURGICAL SOCIETY—"*The Galvanizing of Sheets and Strip by the Hot-Dip Process*," by H. Edwards—Royal Institution of South Wales, Swansea, 6.30 P.M.

15th January

STAFFORDSHIRE IRON AND STEEL INSTITUTE—Joint Meeting with the Iron and Steel Engineers Group, the Iron and Steel Institute, "*Modern Small Rolling-Mills*," by G. A. Phipps—Victoria Hotel, Wolverhampton, 7.0 P.M.

19th January

SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS—Joint Meeting with the Iron and Steel Engineers Group, the Iron and Steel Institute, "*Mechanical and Electrical Features of Primary Hot-Rolling-Mill Appliances*," by W. W. Franklin and P. F. Grove—Royal Victoria Station Hotel, Sheffield, 6.15 P.M.

20th January

CHEMICAL ENGINEERING GROUP OF THE SOCIETY OF CHEMICAL INDUSTRY—"*Abrasion, Erosion, and Corrosion*," by Dr. C. H. Desch, F.R.S.—

XUM

Geological Society, Burlington House, W.1, 5.30 P.M.

22nd January

MIDLAND METALLURGICAL SOCIETIES—"*Protective Coatings*," by W. E. Ballard—James Watt Institute, Great Charles Street, Birmingham, 6.30 P.M.

23rd January

SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS—Joint Meeting with the Institute of Metals (Sheffield Section)—"*Some Metallurgical Problems in the Field of Atomic Energy*," by E. W. Colbeck—Royal Victoria Station Hotel, Sheffield, 6.15 P.M.

TRANSLATION SERVICE

Since the announcement made in the November issue of the *Journal* (*see* page 460), further translations have been put in hand and the following are now available or in course of preparation.

TRANSLATIONS AVAILABLE

No. 326 (German). W. EILENDER and W. ROESER: "Metallurgical Investigations on the Working of the Basic-Bessemer Process with Oxygen-Enriched Blast." (*Stahl und Eisen*, 1939, vol. 59, Sept. 21, pp. 1057–1067).

No. 327 (German). K. KINTZINGER: "Operating Measures in the Sintering of Iron Ores." (*Stahl und Eisen*, 1943, vol. 63, June 10, pp. 453–456). (Slightly abridged).

No. 328 (French). P. LAURENT: "Contribution to the Theory of the Plasticity of Metals and Alloys." (*Revue de Métallurgie*, 1945, vol. 42, Mar., pp. 72–92; Apr., pp. 125–132; May, pp. 156–167; June, pp. 194–203; July, pp. 230–239).

TRANSLATIONS IN COURSE OF PREPARATION

(German). H. BABLIK and F. GÖTZL: "Alloy Additions to Hot-Galvanizing Baths." (*Korrosion und Metallschutz*, 1941, vol. 17, Apr. pp. 128–130).

(German). H. BABLIK, F. GÖTZL and R. KUKACZKA: "The Structure of Hot-Galvanized Coatings." (*Korrosion und Metallschutz*, 1945, vol. 21, No. 1–2, pp. 1–8).

(German). W. BADING: "The Development of Basic Converter Practice." (*Stahl und Eisen*, 1947, vol. 66–67, Apr. 24, pp. 137–149; May 22, pp. 180–186; June 19, pp. 212–223).

(German). W. BAUKLOH: "Desulphurization of Iron by Hydrogen." (*Metallwirtschaft*, 1936, vol. 15, Dec. 11, pp. 1193–1196).

(Swedish). E. BERNHULT: "The Yield Point in Bending and the Stress Distribution in Bars and Tubes in Plastic Bending." (*Jernkontorets Annaler*, 1943, vol. 127, No. 10, pp. 491–533).

(Swedish). E. BERNHULT: "The Calculation of Roll Pressure and Its Effect in the Cold-Straightening of Bars and Tubes." (*Jernkontorets Annaler*, 1944, vol. 128, No. 4, pp. 137–160).

(German). F. BLEILÖB: "Equations for the Time-Elongation Curves in Creep Tests." (*Archiv für das Eisenhüttenwesen*, 1940, vol. 13, May, pp. 489–496).

(German). G. BULLE: "The Production of Pig Iron with Lean Burdens." (*Stahl und Eisen*, 1947, vol. 66–67, Mar. 27, pp. 69–78).

(German). A. EICHINGER: "Mathematical Treatment of Creep Curves." (*Archiv für das Eisenhüttenwesen*, 1940, vol. 13, Mar., pp. 397–402).

(German). A. GELEJI: "The Theoretical Problems Involved in the Design of Rolling-Mill Stands." (*Royal Hungarian Palatine-Joseph University, Publications of the Department of Mining and Metallurgy*, 1941, vol. 13, pp. 224–242).

(German). A. GELEJI: "Calculating the Forces Arising and the Power Requirements in the Mannesmann Tube-Rolling Process." (*Royal Hungarian Palatine-Joseph University, Publications of the Department of Mining and Metallurgy*, 1941, vol. 13, pp. 208–223).

(Swedish). G. MALMBERG: "The Effect of Specimen Shape on the Elongation in Tensile Testing." (*Jernkontorets Annaler*, 1944, vol. 128, No. 6, pp. 197–245).

(Swedish). S. MÖRTSELL: "Rationalization in Swedish Iron Ore Dressing." (*Jernkontorets Annaler*, 1946, vol. 130, No. 9, pp. 369–460).

(German). W. OELSEN and H. WIEMER: "Immiscibility Phenomena in Ferrous-Oxide/Sodium-Phosphate Slags." (*Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung*, 1942, vol. 24, No. 13, pp. 167–210).

(German). A. POMP and W. PUZICHA: "The Cold-Rolling of High-Tensile Strip Steels and Their Properties." (*Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung*, 1943, vol. 26, No. 2, pp. 13–36).

(German). E. SIEBEL: "Present Knowledge of the Mechanics of Wire Drawing." (*Stahl und Eisen*, 1947, vol. 66–67, May 22, pp. 171–180).

(Russian). A. N. TIKHONOV and E. G. SHOIDOVSKI: "Mathematical Theory of Continuous Casting." (*Journal of Technical Physics, U.S.S.R.*, 1947, vol. 17, No. 2, pp. 161–176).

(Russian). F. F. VITMAN, N. N. DAVIDENKOV, and N. A. ZLATIN: "Cold-Brittleness of Steel in Torsion." (*Journal of Experimental and Theoretical Physics, U.S.S.R.*, 1940, vol. 10, Nos. 9–10, pp. 1137–1145).

CHARGES FOR COPIES OF TRANSLATIONS.—For the above translations a charge will be made of 10*s*. for the first copy and 5*s*. each for each additional copy of the same translation. Requests should be accompanied by a remittance. The above translations are not available on loan from the Joint Library.

TRANSLATIONS PREPARED AT MEMBERS' REQUEST.—Members requiring translations of foreign papers are invited to communicate with the Secretary, who will ascertain whether the translations can be prepared for inclusion in the series.

ABSTRACTS OF CURRENT LITERATURE *and* BOOK NOTICES

IRON AND STEEL MANUFACTURE AND RELATED SUBJECTS

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MINERAL RESOURCES

The Iron Mine of Couthuin in the Province of Liége. C. Ancion, J. Tordeur and W. Van Leckwijck. (Revue Universelle des Mines, 1946, Series 9, vol. 2, No. 3, pp. 118–128). A description is given of the Couthuin mine, in which a palaeozoic oolitic ore is worked.—R. F. F.

Solution of the Problem of the Kursk Magnetic Anomaly. V. Rikman. (Metallurgia, 1947, vol. 36, Aug., pp. 201–202). A short summary is given of the investigations which led to the discovery of vast iron-ore resources in the Kursk region. It is estimated that the reserves amount to about 340 million tons of rich ore (up to 70% of iron) and over 200,000 million tons of poorer quartzites. The exploitation is difficult because of water-bearing sands which overlie the ore, but plans have been made for the establishment of large iron and steel works in the district.—C. O.

Iron Ore Supply for the Future. G. W. Hewitt. (American Iron and Steel Institute: Blast Furnace and Steel Plant, 1947, vol. 35, Aug., pp. 957–963). A detailed study is made of the iron-ore reserves of the United States, and statistics of shipments and smelting costs are given. It is expected that the theoretical 20–25 years' supply of first-grade Lake Superior ores will to a great extent be supplemented by the production of high-grade concentrates from the intermediate ores and the taconites.—J. R.

The World's Manganese Supply. R. Durrer. (Von Roll Werkzeitung, 1947, vol. 18, Sept., p. 137). Statistics on the world's resources, production, and consumption of manganese are presented and discussed.—R. A. R.

The Most Important Ore Resources in the Russian Occupation Zone of Germany. A. Reimers. (Technik, 1947, vol. 2, Aug., pp. 361–365).

The Union of Iron Ore Mines and Its Tasks. T. Schrotter. (Hutnik, 1946, vol. 13, Nov., pp. 552–562). [In Polish]. The types and quantities of ore available in the various Polish mines are discussed in relation to the three-year plan for increased output.—W. J. W.

The World Production of Iron Before and During the Second World War. I. Borejdo. (Hutnik, 1946, vol. 13, July, pp. 341–351; Aug., pp. 409, 425; Sept., pp. 457–462). [In Polish]. Details of iron production are given, and resources of ores and plant in the United States, the U.S.S.R., and the United Kingdom are discussed.—W. J. W.

Present Knowledge of Geological Deposits. K. Bohdanowics. (Hutnik, 1946, vol. 13, Mar., pp. 113–119; Apr., pp. 180–184; May, pp. 264–269; June, pp. 229–305). [In Polish]. This is a general discussion of geological deposits throughout the world.—W. J. W.

FUEL—PREPARATION, PROPERTIES, AND USES

Relation of the Power Load to the Steam and Fuel Balances. H. L. Halstead. (Iron and Steel Engineer, 1947, vol. 24, Aug., pp. 38–43). The calculation of the fuel balance, steam requirements, and electrical generating capacity of a hypothetical new integrated iron and steel works is illustrated by numerical examples.—C. O.

Power, Fuel and Steam Balances. A. D. Howry. (Iron and Steel Engineer, 1947, vol. 24, Aug., pp. 35–38). A completely balanced electrical power system for a hypothetical new plant containing blast-furnaces, coke-ovens, steel furnaces, and rolling mills is outlined. —C. O.

The Generation and Utilization of Steam in the Heavy Industries. G. E. Hider. (Proceedings of the Institution of Mechanical Engineers, 1946, vol. 155, pp. 376–384). A graphical method for the analysis of heat transmission in boilers is developed and its application to problems such as waste-heat boilers in steelworks is explained.—C. O.

Characteristics of Basic Gas Combustion Equipment. D. A. Campbell. (Steel, 1947, vol. 121, Sept. 22, pp. 78–79, 103–108). Gas combustion systems are classified into six groups, according to the method by which the gases are mixed, and the characteristics of each are discussed. A summary is made of the common terms and formulæ used in combustion engineering.—C. O.

The Recovery of Industrial Low-Grade Heat. W. A. MacFarlane and J. B. Mason. (Fuel Economy Conference, World Power Conference, The Hague, 1947, Section C2, Paper No. 10). The principal sources from which industrial waste heat may be reclaimed are discussed and illustrated by descriptions of the economies effected by modifications to the layout of a dairy, a brewery, and a paper-mill.—C. O.

Electronic Control of Combustion in Steam Boilers. Z. Ryska. (Fuel Economy Conference, World Power Conference, The Hague, 1947, Section C2, Paper No. 12). The principles of electronic control of the firing of boilers are illustrated by a detailed description of the method of regulation of a travelling-grate stoker.—C. O.

New Ways of Burning Liquid Fuel. W. L. v. d. Putte and H. J. K. v. d. Bussche. (Fuel Economy Conference, World Power Conference, The Hague, 1947, Section C2, Paper No. 16). A detailed description is given of the development of a completely automatic rotary-cup type of oil burner suitable for use in domestic installations. The burner gives fine atomization and can operate on fuels of widely different types.—C. O.

Fuel Requirements in Steelmaking. M. W. Thring. (British Association : Coke and Gas, 1947, vol. 9, Oct., pp. 299–306). The coal consumption in the various steps of the production of pig iron and steel is analysed, and the disparity between theory and practice is indicated. The author summarizes investigations recently completed or now in progress on the instrumentation of furnaces and the study of the rate-determining factors in furnace reactions. Recent developments leading to increased economy in fuel consumption which are mentioned include the Krupp-Renn process, oxygen-enriched blast, high top pressures in blast-furnaces, and the use of oxygen in the open-hearth furnace and the Bessemer converter.—C. O.

New Gas or Gas-Oil Burner. (Iron and Coal Trades Review, 1947, vol. 155, Sept. 26, p. 617). An illustrated description is given of a new gas or gas-oil burner. The chief features of the new burner are a high standard of efficiency and extreme flexibility, and it is claimed that any industrial gas, whether clean and cold, or hot and dirty, can be used without modification to the burner design.—J. R.

The Gas Turbine as a Means of Securing Additional Output of Power. W. Karrer. (Fuel Economy Conference, World Power Conference, The Hague, 1947, Section C2, Paper No. 13). It is shown how the air turbine may be embodied in steam-power plants to secure additional power with a low fuel consumption. The increased output and the efficiency with which it can be obtained are considerably higher than are realized with any type of pure steam-power plant.—C. O.

A New Metal Recuperator for High Temperature Furnaces. H. Escher. (Fuel Economy Conference, World Power Conference, The Hague, 1947, Section C2, Paper No. 9). A type of metallic recuperator is described which is suitable for use with high-temperature gases containing high concentrations of dust. The application of a recuperator of this design to a 125-ton open-hearth steel furnace is discussed.—C. O.

The Origin and Movement of Sulphur in Coal. W. Petrascheck. (Berg- und Hüttenmännische Monatshefte der Montanistischen Hochschule in Leoben, 1947, vol. 92, June, pp. 104–106).

The Measurement of Viscosity of Ash of Some Polish Coals at High Temperatures. K. Ackerman. (Hutnik, 1946, vol. 13, Oct., pp. 477–486). [In Polish].

Pulverized Fuel. T. F. Hurley. (Nature, 1947, vol. 160, Sept. 6, pp. 318–320). A short account is given of the rapidly increasing use of pulverized coal as an industrial fuel, and of the control and removal of ash, which constitutes the main obstacle to the extended application of pulverized-coal firing.—C. O.

Apparatus for Measuring the Flow of a Mixture of Air and Pulverized Coal. A. Jorissen. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 9, pp. 349–353). An American apparatus for measuring the flow of mixed air and pulverized coal, consisting of a rubber orifice plate and a nozzle, the first giving the air flow and the second the flow of the mixture, is discussed.—R. F. F.

Coke Ovens with "Free Expansion." (Génie Civil, 1947, vol. 124, Oct. 15, pp. 396–397). The construction of coke-ovens, using silica bricks only for the chamber, and with the necessary allowance for the expansion of silica bricks, is described.—R. F. F.

Preparing a Coke-Oven Slack. (Coke and Gas, 1947, vol. 9, Oct., pp. 313–317). Details are given of the washery at the coal-blending plant at the Redbourn works of Messrs. Richard Thomas and Baldwins, Ltd. Up to 125 tons per hour of ½-in. slack are passed through a standard Coppée Baum washer and the − ½-mm. slurry is treated in a vacuum froth-flotation system.—C. O.

Hydrogen Sulphide Removal by Ammoniacal Iron Ammonium Ferrocyanide Liquors. H. C. Craggs and M. H. M. Arnold. (Chemistry and Industry, 1947, Sept. 20, pp. 571–573; Sept. 27, pp. 590–593). The efforts made at the Billingham plant of Imperial Chemical Industries, Ltd., to improve the autopurification process for the removal of hydrogen sulphide from carbonization gases are described.—C. O.

New Practical Method for Determining the Density of Coke. G. Bandelet. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Dec. 15, pp. 208–212). A method is given for determining the density of coke.—R. F. F.

Fuel Economy in Coal Carbonisation. T. C. Finlayson. (Fuel Economy Conference of the World Power Conference, The Hague, 1947 : Gas World, 1947, vol. 127, Sept. 6, pp. 321–327). Recent investigations and reports bearing on the subject of fuel economy in the gas and coke-oven industries are reviewed.—C. O.

Study on the Caking Properties of Coals by the "Meurice" Method. A. Benoit. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 141–144). Tests carried out at the laboratory of the Cokerie d'Ougrée-Marihaye on the "Meurice" method, show the need for standardizing the operations of this method.—R. F. F.

Measurement of the Various Characteristics of the Plastic Range of Coal. A. Gillet. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 144–154). The various tests to determine the coking value of coals are reviewed and compared.—R. F. F.

Gas Utilisation at a Steelworks. (Coke and Gas, 1947, vol. 9, Aug., pp. 225–229). Details are given of the reorganization of the boilers and gas supplies at the

Workington plant of the United Steel Companies, Ltd., whereby the whole of the steam requirements are met from the coke-oven and blast-furnace gas produced in the works.—C. O.

Fuel Oil in the Steel Industry. (British Iron and Steel Federation, Monthly Statistical Bulletin, 1947, vol. 22, Aug., No. 8). The technical and economic aspects of the use of fuel oil in steel manufacture, particularly for firing open-hearth furnaces, are discussed.—M. A. V.

TEMPERATURE MEASUREMENT AND CONTROL

Temperature Measurements in Steel Baths. S. Fornander. (Jernkontorets Annaler, 1947, vol. 131, No. 7, pp. 225–242). [In Swedish]. The quick-immersion-thermocouple method of determining the temperatures of liquid steel was investigated. A pyrometer of British design for high-frequency induction furnaces gave good results with respect to reproducibility and accuracy. As many as 80 determinations could be made with a single hot junction without any measurable change in e.m.f. A pyrometer based on British types was made for open-hearth and electric furnaces, but the construction was modified so that no parts were likely to give off gases when heated. Good results were obtained and up to 50 immersions could be made without change in the e.m.f.—R. A. R.

Measurement of Heating Rates in Reheating Furnaces. J. W. Percy. (Iron and Steel Engineer, 1947, vol. 24, July, pp. 65–69). A technique evolved to measure the temperature at the centre of slabs, 4 to 7 in. thick, during their passage through a pusher-type reheating furnace is described. A 1-in. dia. hole was bored along the major axis to the geometric centre of the slabs to be tested, the slabs being arranged in the furnace in such a manner that a platinum/platinum-rhodium thermocouple 10 ft. long, enclosed in a $\frac{3}{8}$-in. stainless-steel tube, could be inserted in the hole as the slabs passed each of the doors along the sides of the furnace. The rate of "pushing" was adjusted to allow the slab to remain in front of each door for the two minutes necessary for the temperature measurements to be made with an accuracy of $\pm$ 5° F. Surface temperature measurements were also made by means of a water-jacketed radiation pyrometer resting on the surface of the slabs and requiring an exposure of only 10–15 sec.—C. O.

Selection and Installation of Thermocouple Leadwires. (Steel, 1947, vol. 121, Aug. 18, pp. 91, 112–118). The factors to consider when selecting and installing the lead wires in industrial thermocouple circuits are discussed.—C. O.

Fundamentals and a Few Technical Terms Dealing with Automatic Temperature Control for Industrial Purposes. N. Retlaw. (Instrument Practice, 1947, vol. 1, Aug., pp. 381–386). The principles of automatic temperature control and a number of technical terms used in its application are briefly explained.—C. O.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Electric Smelting. R. Durrer. (Stahl und Eisen, 1947, vol. 66–67, Sept. 11, pp. 312–315). This is a German translation of a paper which appeared in Journ. I. and S.I., 1947, vol. 156, June, pp. 257–260.—R. A. R.

A Plan for Rebuilding the Iron Industry. S. Wroblewski. (Hutnik, 1946, vol. 13, Nov., pp. 525–551). [In Polish]. The Polish iron industry is reviewed generally and details given of the plant available in 1938 and also in the recently regained part of Poland. The loss of plant during the occupation of Poland is discussed and details given of a three-year plan for increasing output.—W. J. W.

An Introduction to the Interaction of Carbon and Iron Ore at Temperatures up to 1450° C. H. L. Saunders and H. J. Tress. (Journal of the Iron and Steel Institute, 1947, vol. 157, Oct., pp. 215–222). The reduction of iron ore in the blast-furnace proceeds at all levels from the stack to the hearth, but the mechanism changes with increasing temperature. In the stack the reaction is "indirect," the CO of the reducing gas being converted to CO_2, whilst in the hearth the carbon reacts "directly" to give CO. Gaseous reduction diminishes as the porosity of the ore is reduced by the partial fusion of the gangue, the remaining oxygen being eliminated by reactions between ferrous oxide and various forms of carbon present at high temperatures. A number of comparative reduction rates have been determined from three ores and several types of carbon.

The Distribution of Temperature in the Blast-Furnace and Its Effect on the Blast-Furnace Process. W. Kuczewski. (Hutnik, 1942, vol, 13, Apr., pp. 184–204). [In Polish]. Blast, slag, and metal temperatures in blast-furnaces operating on various burdens are given. It is suggested that in addition to normal methods of blast-furnace control, a knowledge of the changes in temperature in a vertical section of the furnace is important.—W. J. W.

The Production of Iron and Steel with Oxygen-Enriched Blast. R. Durrer. (Stahl und Eisen, 1947, vol. 66–67, July 17, pp. 239–241). This is a German translation of the paper which appeared in Journ. I. and S.I., 1947, vol. 156, June, pp. 253–256—R. A. R.

Results of Operating Blast Furnaces with an Acid Burden. A. Portevin. (Bulletin de la Société d'Encouragement pour l'Industrie Nationale, 1940, vol. 139, May–June, pp. 150–152). The Brassert acid-burden process and its application at Corby are reviewed, and the possibility of its use with Normandy ores is discussed briefly.—J. C. R.

Economics of the Blast Furnace. B. S. Old, A. R. Almeida, R. W. Hyde, and E. L. Pepper. (Iron Age, 1947, vol. 160, Sept. 18, pp. 60–69). Recent literature on the use of high top pressures in blast-furnaces is reviewed. Estimates are made from the published data of the results which might be obtained by operating (*a*) at still higher pressures, (*b*) with combined oxygen-enriched blast and high top pressures, and (*c*) with combined ore beneficiation, oxygen-enriched blast, and high top pressures. The calculations indicate that method (*c*) should enable the furnace output to be doubled.—C.O.

The Use of Carbon in the Blast Furnace and Heat Balances. T. L. Joseph and K. Neustaetter. (Blast Furnace and Steel Plant, 1947, vol. 35, July, pp. 824–831; Aug., pp. 944–948). The reactions involving the consumption of carbon in the blast-furnace and their relationship to the efficiency of operation are discussed. An example is given of the calculation of a heat balance, and the article concludes with a discussion of the effect of blowing rates upon solution loss.—J. R.

Aluminium Tuyeres. R. Guillemot. (Revue de l'Aluminium, 1946, vol. 23, Apr., pp. 137–141). The history of the use of aluminium tuyeres in blast-furnaces is reviewed, with particular reference to their application in French steelworks.—J. C. R.

TREATMENT AND USE OF SLAGS

Recent Experience in the Use of Slag-Lime as a Fertilizer. H. Kappen. (Stahl und Eisen, 1947, vol. 66–67, Sept. 11, pp. 307–312).

Utilization of Sydney Slags for Engineering and Agriculture. N. A. Parlee. (Engineering Journal, 1947, vol. 30, Jan., pp. 14–16). The use of blast-furnace slag from the Sydney plant of the Dominion Steel and Coal Corporation for railway ballast, road building, insulation, and in making light concrete, and of open-hearth slags as fertilizers is discussed.—R. A. R.

Report on Investigations in Germany on Slag Cements and Lightweight Concrete. (British Intelligence Objectives Sub-Committee, 1947, Final Report No. 1131: H.M. Stationery Office).

PRODUCTION OF STEEL

Modernisation of Iron and Steel Industry—Samuel Fox & Company Limited. (Engineer, 1947, vol. 184, July 4, pp. 8–9; July 11, pp. 30–31; July 18, pp. 56–57). A comprehensive review is made of the modernization schemes completed or in hand at the chief British iron and steel works.

Modernisation of the Iron and Steel Industry. (Man and Metal, 1947, vol. 24, Oct., p. 152.). Developments which have either been recently completed or are now in hand are described. These include the conversion of three 75-ton basic open-hearth furnaces to oil firing, and the installation of a number of new units for strip production.—M. A. V.

Modernisation of the Iron and Steel Industry—2. (Man and Metal, 1947, vol. 24, Sept., pp. 134–135). Recent developments at the plant of the Workington Iron and Steel Company and proposed extensions and reconstruction at the Appleby-Frodingham Works of the United Steel Companies are described. The chief developments envisaged at the latter works are the reconstruction and electrification of the section rolling mills and the replacement of the present Frodingham melting shop by a new melting shop now under construction.—M. A. V.

Reconstruction by the Steel Company of Wales. (Iron and Coal Trades Review, 1947, vol. 155, Aug. 29, pp. 419–420). A summary is made of the construction schemes being carried out at the Margam, Abbey, and Port Talbot works of the Steel Company of Wales, Ltd. —C. O.

Factors which Determine Iron and Steel Making Processes. H. W. Graham. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2217: Metals Technology, 1947, vol. 14, Aug.). The essential bases of the technology of iron- and steel-making are examined at some length, and the interrelation of processes and the factors which determine their use or neglect are discussed. It is considered that the end of steelmaking by melting methods cannot yet be envisaged.—C. O.

The Gas Balance in Iron and Steel Works. K. Rummel. (Iron and Steel Institute, 1947, Translation Series, No. 323). An English translation is presented of a paper which appeared in Stahl und Eisen, 1947, vol. 66–67, Jan. 2, pp. 19–23 (*see* Journ. I and S.I., 1947, vol. 155, Mar., p. 455).—R. A. R.

Remarks on the American Iron and Steel Industry. P. Coheur. (Revue Universelle des Mines, 1946, Series 9, vol. 2, No. 6, pp. 229–234). Recent progress in iron and steel manufacture in America is reviewed.—R. F. F.

Skoda Works, Pilsen, Czecho-Slovakia. (British Intelligence Objectives Sub-Committee, 1947, B.I.O.S. Final Report No. 43: H.M. Stationery Office). A very short general account is given of German production methods at the Pilsen armaments works.—C. O.

Reactions in the Acid Side Blown Converter. N. F. Dufty. (Metallurgia, 1947, vol. 36, Aug., pp. 179–181). An attempt is made to develop theories for the reactions in the acid side-blown converter process on the basis of the data published in recent literature.—C. O.

Manufacture of Basic Bessemer Steel Low in Nitrogen. Y. Dardel. (Metal Progress, 1947, vol. 52, Aug., pp. 252–256). Various methods which have been used to reduce the nitrogen content of basic-Bessemer steels are discussed. It seems possible to obtain contents as low as 0·001–0·002% by blowing with pure oxygen, but this is very expensive. The author suggests that the nitrogen may be reduced to 0·003–0·005% by blowing with air during the first 9 or 10 min. and using pure oxygen for the afterblow.—C. O.

Practical Conclusions Resulting from a Theory of Heat Transfer in the Open-Hearth Furnace. M. W. Thring. (Iron and Coal Trades Review, 1947, vol. 155, Oct. 3, pp. 647–652). Open-hearth-furnace operation is discussed in the light of a heat-transfer theory in which the method of successive approximations is used.—J. R.

The Mechanism of Carbon Removal in the Open-Hearth Furnace. A. H. Jay. (Journal of the Iron and Steel Institute, 1947, vol. 157, Oct., pp. 167–172). An account is given of the development of a suggested mechanism for the carbon-removal action from liquid steel in open-hearth furnaces.

A critical examination of a typical acid open-hearth cast shows that the rates of carbon removal at two definite periods (*i.e.*, at the end of the ore boil and during the limestone boil, when the respective carbon contents are widely dissimilar) are closely related only to the degrees of oxidation at these stages (as judged by the respective silicon and manganese contents) rather than to the product of the carbon and oxygen contents. This conclusion is supported by many previous observations of steelmaking data. On this basis, a mechanism is evolved which postulates the following chain of reactions: (1) C (Soluble) + O (soluble) $\longrightarrow$ CO (insoluble molecular gas bubble); (2) $CO + O \longrightarrow CO_2$; (3) $CO_2 + C \longrightarrow 2CO$. A two-stage reaction (equations (2) and (3)) thus proceeds with subsequent growth of the bubble. In view of the very low ratio of oxygen to carbon in the medium-high carbon ranges, it is considered that reaction (2) is the governing factor in determining the ultimate speed of carbon removal.

The hypothesis is also substantiated by calculations of the relative concentrations of oxygen and iron atoms on the surface of the primary molecular CO gas bubble. These show that the CO gas bubble can accommodate 3300 iron atoms. If, therefore, at least one oxygen atom is present on the bubble surface, as might well be necessary in order to allow the reaction to proceed, it can readily be shown that the overall oxygen content of the liquid bath would be 0·009%. This value is in remarkable agreement with experimental data—particularly American—which indicate that for steady carbon removal a minimum excess oxygen content, *e.g.*, 0·009% (*i.e.*, excess oxygen over the equilibrium value) is required in the medium–high carbon ranges.

Factors Affecting Basic Open-Hearth Operating Rates. R. H. Ede. (Steel, 1947, vol. 121, Sept. 8, pp. 94–105). The use of the statistical method of multiple correlation in determining the factors which govern the rate of operation of a basic open-hearth furnace is explained.—C. O.

Economics of Oxygen Use in Steel Making Furnaces. Linde Air Products Company. (Blast Furnace and Steel Plant, 1947, vol. 35, July, pp. 813–816). A brief indication is made of the way in which the use of oxygen would influence each of the factors which contribute to the cost of open-hearth steelmaking. It is considered that in general, savings may be obtained by scrap-melting

and carbon reduction with the oxygen jet, but that oxygen supplied through end burners gives increased production only at increased cost.—C. O.

Oxygen and the Open Hearth Furnace. (Canadian Metals and Metallurgical Industries, 1947, vol. 10, July, pp. 16–22). Details are given of the shortening of the melt-down period in open-hearth steelmaking brought about by : (*a*) Supplying oxygen with the fuel in the main burners ; (*b*) melting preheated scrap by the use of a high-velocity oxygen jet ; (*c*) supplying oxygen through auxiliary burners ; and (*d*) enriching the combustion air in the checkers with oxygen. The figures quoted were obtained in laboratory and experimental furnaces.—C. O.

New Method of Ramming Steel Furnace Hearths. M. Stankiewicz. (Hutnik, 1946, vol. 13, Mar., pp. 150–154). [In Polish]. A new method of repairing hearths of open-hearth and electric furnaces is described and details given of the materials required.—W. J. W.

Electrolytic Manganese in Stainless Steel. F. Sillers, jun., and R. T. C. Rasmussen. United States Department of the Interior 1947, June, Bureau of Mines Report of Investigation 4078). Data are given on 81 heats of five types of stainless steel made with electrolytic manganese in place of ferromanganese. The substitution was entirely satisfactory from the points of view of steelmaking and the properties of the steel produced, but the cost of electrolytic manganese is considered to make its use uneconomical.—C. O.

Proceedings of the Fourth Electric Furnace Steel Conference. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, 1946, vol. 4). This Conference was held on December 5, 6, and 7, 1946, in Pittsburgh, and was divided into nine sessions, dealing with (*a*) electric-furnace electrodes, (*b*) the metallurgy of acid and basic steels, (*c*) charging, (*d*) oxidizing-slag heats, (*e*) stainless steel, (*f*) high-speed steel, (*g*) the melting of high-alloy steels, (*h*) the importance of slag in acid practice, and (*i*) sulphur, its effect and control. Abstracts of the papers follow.—C. O.

Carbon and Graphite Electric-Furnace Electrodes. T. L. Nelson. (*Ibid.*, pp. 5–13). A résumé is made of the nature and properties of graphite and carbon steel-furnace electrodes. Particular reference is made to the design and testing of an Acme-threaded graphite nipple.—C. O.

The Mechanism of the Carbon-Oxygen Reaction in Steelmaking. C. E. Sims. (*Ibid.*, pp. 15–30). *See* Journ. I. and S.I., 1947, vol. 156, May, p. 132.

Oxygen in Basic Electric-Furnace Baths. S. F. Urban and G. Derge. (*Ibid.*, pp. 31–49). *See* Journ. I. and S.I., 1947, vol. 157, Oct., p. 291.

The Origin of Silicate Inclusions in Basic Electric-Arc-Furnace Steel of Higher Carbon Contents. A. Hultgren. (*Ibid.*, pp. 50–80). This paper was orginally published in Jerkontorets Annaler, 1945, vol. 129, No. 11, pp. 633–671. (*See* Journ. I. and S.I., 1946, No. I, p. 42A).

Preparation of Raw Materials for Fast Melting in an Electric-Arc Furnace. S. D. Gladding. (*Ibid.*, pp. 81–87). *See* Journ. I. and S.I., 1947, vol. 156, May, p. 132.

The Oxidizing Period in Basic Electric Steelmaking. A. C. Texter. (*Ibid.*, pp. 87–104). The purpose of oxidation in basic electric steelmaking, and its influence on quality are discussed. Details are given of the oxidizing practices followed by Atlas Steels, Ltd., Welland, Ontario, in 35-ton furnaces.—C. O.

Melting Semikilled Steel in the Basic Electric Furnace. R. J. McCurdy and R. W. Farley. (*Ibid.*, pp. 105–107). The essential details are given of a single-slag practice for the rapid melting of semi-killed steel in basic electric-arc furnaces. The heat is tapped without blocking, all additions are made in the ladle, and open-top moulds are used.—C. O.

Production of Rimmed Steel in the Basic Electric Furnace. A. K. Moore. (*Ibid.*, pp. 107–110). A short account is given of the techniques used by the Steel Company of Canada for the production of all types of rimming steels in the basic electric furnace.—C. O.

Steelmaking Operations at Connors Steel Company. B. C. Blake. (*Ibid.*, pp. 110–111). A brief description is given of the electric-steel-furnace equipment and practice of the Connors Steel Company, Birmingham, Alabama. Two top-charge Lectromelt basic electric furnaces are used to melt plain-carbon steels.—C. O.

Practice by Burlington Steel Company Limited. G. H. McCally. (*Ibid.*, pp. 112–113). A short account is given of the slag-control methods used by the Burlington Steel Company, Ltd., Hamilton, Ontario, in the manufacture of straight carbon steel in a basic Lectromelt furnace.—C. O.

Melting Stainless Steel at South Works. J. H. Eisaman. (*Ibid.*, pp. 123–129). Details of the stainless-steel melting practice at the South Works of the Carnegie-Illinois Steel Corporation, Chicago, Illinois, are given. Decarburization is carried out in the 80-ton Héroult furnaces by means of oxygen directed through lances on to the metal/slag interface.—C. O.

Economics of Stainless-Steel Production. C. Staley. (*Ibid.*, pp. 129–133). A brief account is given of the utilization of scrap in making stainless steel at the Middletown Works, Ohio, of the American Rolling Mill Company.—C. O.

Possibility of Use of Grinding Dust in Making Stainless Steel. M. J. Meinen. (*Ibid.*, pp. 126–137). In a brief account of the practice of the Crucible Steel Co. of America, Pittsburgh, Pa., in melting stainless steel, mention is made of experiments which have been made with the use of dust from the grinding of stainless steel as a substitute for scrap. Under suitable conditions about 70% of the metal in the dust can be recovered.—C. O.

Use of Inert Gas. P. M. Hulme. (*Ibid.*, pp. 137–140). A short review is made of progress in the year 1946 in the removal of dissolved gas from molten steel by the use of argon and nitrogen.—C. O.

Progress in the Use of Argon Gas. W. M. Farnsworth. (*Ibid.*, pp. 140–145). Details are given of a method of controlling "bleeding" in electric-furnace steel by bubbling argon through the bath.—C. O.

Melting of High-Speed Steel in the Electric-Arc Furnace. R. Schemp and J. B. Schrader. (*Ibid.*, pp. 146–154). A general review is made of modern practice in the melting of high-speed steel in the electric-arc furnace.—C. O.

Arc *versus* Induction-Furnace Melting. S. F. Urban. (*Ibid.*, pp. 159–161). The merits of the arc and induction furnaces for melting stainless steel are compared—the author favouring the latter.—C. O.

Notes on Melting Heat and Corrosion-Resistant Metals by the Acid Electric Process. F. O. Lemmon. (*Ibid.*, pp. 163–170). The melting of nickel-chromium heat- and corrosion-resistant high-alloy steels is briefly discussed.—C. O.

Induction Melting of High-Alloy Steels. R. J. Wilcox. (*Ibid.*, pp. 170–175). An outline is given of the practice of the Michigan Steel Casting Co., Detroit, Michigan, in melting high-alloy steels in 1000-lb. and 100-lb. capacity Ajax-Northrup high-frequency induction furnaces.—C. O.

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Acid Electric Slags. G. R. Fitterer. (*Ibid.*, pp. 185–196). The literature on the slag-fluidity test is reviewed, and the results of tests on the slags from six heats of acid electric steel are presented. It is concluded that slag fluidity may be used to indicate changes in the following factors : (1) Slag analysis ; (2) metal temperature ; (3) rate of carbon drop ; and (4) finishing conditions.—C. O.

Acid Electric Slags Are a Necessary Evil. E. C. Troy. (*Ibid.*, pp. 199–204). After a short account of the various methods of classifying slags, the author discusses the functions of the slag in acid electric steelmaking practice.—C. O.

You're Stuck with Slag—Why Not Use It ? J. B. Caine. (*Ibid.*, pp. 204–209). The advantages which may be obtained by utilizing the facts that the free iron oxide content in acid slags is relatively unaffected by other slag components, and that it is much less than in basic slags, are indicated.—C. O.

The Perrin Process. J. Strauss. (*Ibid.*, pp. 236–238). A short outline is given of the Perrin steelmaking process.—C. O.

Desulphurized Steel Made by the Triplex Process. H. BeMent. (*Ibid.*, pp. 241–243). A short account is given of the use of soda-ash for desulphurizing steel in the ladle between the cupola and converter phases of the triplex process.—C. O.

Possibilities of Use of Spodumene for Removal of Sulphur. G. T. Motock. (*Ibid.*, pp. 245–246). A very brief account is given of tests which have been made with spodumene ($Li_2O.Al_2O_3.4SiO_2$) for the desulphurization of steel. Sulphur removals of 55% from the steel, and 45% from the slag, possibly by the volatilization of lithium sulphides, are indicated.—C. O.

Single Slag Basic Electric Heats. M. V. Healey. (Steel, 1947, vol. 121, Aug. 25, pp. 104–105). A basic electric furnace practice is discussed by means of which it is possible to produce molybdenum and carbon steels with grain-growth characteristics similar to those of basic open-hearth steels. The bath is boiled under a single oxidizing slag for a period sufficient to ensure the elimination of all neutral and reducing gases, and it is tapped before the gases are able to return.—C. O.

The Rapid Determination of the Basicity of Basic Open-Hearth Slags from Their Mineral Origin. F. Trojer. (Berg- und Hüttenmännische Monatshefte der Montanistischen Hochschule in Leoben, 1947, vol. 92, Jan.–Mar., pp. 35–49).

The Chemistry of Steelmaking Slags. J. E. Garside. (Iron and Coal Trades Review, 1947, vol. 155, Sept. 12, pp. 497–501 ; Sept. 19, pp. 553–557 ; Sept. 26, pp. 603–606). A survey is made of the literature on the physical chemistry of steelmaking slags and refining reactions. —C. O.

The Production of High Speed and Other Special Alloy and Carbon Steels in Germany and Austria. (British Intelligence Objectives Sub-Committee, 1947, Final Report No. 1505 : H.M. Stationery Office). A detailed summary is made of the products, output, equipment, and steelmaking practices of sixteen German and Austrian firms manufacturing special alloy and carbon steels. No confirmation of any successful method of continuous casting was obtained.—C. O.

An Electrochemical Study of the Properties of Molten Slags of the Systems $CaO–SiO_2$ and $CaO–Al_2O_3–SiO_2$. Lo Ching Chang and G. Derge. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2101 : Metals Technology, 1946, vol. 13, Oct.). A method for studying the properties of molten slags by measurements of reversible electrode potentials is described. Of the electrode combinations used in studying the systems $CaO–SiO_2$ and $CaO–Al_2O_3–SiO_2$, the pair silicon-carbide/graphite gave the best results. Data are given on the effect of composition and temperature on the e.m.f. measurements, and equations are suggested for the electrode reactions.—C. O.

Some Mathematical Considerations and Experiments Concerning the Shape of Pipes in Ingots. E. W. Fell. (Journal of the Institute of Metals, 1947, vol. 73, Jan., pp. 243–262). The shapes of the pipes formed during the solidification of finite vertical cylinders of molten metal are determined mathematically when cooling proceeds from (*a*) the vertical sides only, and (*b*) from the vertical sides and the flat bottom. The analysis is compared with results obtained experimentally in casts of aluminium alloys.—C. O.

FOUNDRY PRACTICE

Fifty Years Jubilee of Giesserei Bern. E. Fankhauser. (Von Roll Mitteilungen, 1944, vol. 3, Dec., pp. 101–104). A description is given of the development of the foundry " Giesserei Bern " since it was taken over by Ludw. von Roll'schen Eisenwerke, A.G., in 1894.—R. A. R.

Slagging Gas Producer and Its Uses in Metallurgy. H. Philipon. (Revue de l'Industrie Minérale, 1942, May–June, pp. 93–99). The author describes a type of installation to produce a high quality cast iron with little loss of material. The properties of the cast irons produced are discussed and the use of a slagging gas-producer-cupola is considered from the point of view of gas utilization in a steelworks.—R. F. F.

Does It Pay to Mould in Silica Sand with an Addition of Cement ? J. Vorlíček. (Hutnické Listy, 1947, vol. 1, No. 9, pp. 200–202 ; No. 10, pp. 220–223). [In Czech]. The advantages and disadvantages of making grey-iron castings in sand-cement moulds are discussed.—R. A. R.

Controlling Casting Quality with a Dilatometer. C. M. King. (Iron Age, 1947, vol. 160, July 3, pp. 73–74). The benefits which may be derived from the complete knowledge of the behaviour of a moulding sand at the metal-pouring temperature are summarized. A very short description is given of an instrument known as the " Dilatometer," which enables the properties of cylinders of moulding sands under load at high temperatures to be measured.—C. O.

The Influence of Production Flow on Moulding Methods in Iron Foundries and Its Effect on P.M.H. and General Efficiency. R. C. Shepherd. (Institute of British Foundrymen : Foundry Trade Journal, 1947, vol. 83, Sept. 11, pp. 21–30, 33 ; Sept. 18, pp. 43–51 ; Sept. 25, pp. 67–75). The need for giving immediate consideration to the employment of improved moulding methods is discussed and a greater use of production-flow operation in moulding shops is advocated. Moulding work in general is broadly reviewed and related to appropriate production-flow methods, the effects of which are dealt with. Typical continuous and bank systems of moulding operation are illustrated, and units incorporating combinations of both systems are considered for approximate mixtures of product ; in some cases this involves multiple-operation mould-making, as distinct from the use of a single-purpose machine. The segregation of moulding operations for all classes of work is suggested as a means of increasing the production per man-hour and conserving skilled labour.—J. R.

A New Automatic Moulding Machine. (Foundry Trade Journal, 1947, vol. 83, Sept. 25, pp. 65, 66). A description is given of a new type of moulding machine in which the full cycle of operations automatically

follows in correct sequence on the pressing of a push-button. The logical outcome of this development is battery operation, with opportunity for very marked economy in operation.—J. R.

Conference on Rolls. (Revue Universelle des Mines, 1945, Series 9, Vol. 1, Aug. 15 and Sept. 15). This conference was organized during the winter 1943–1944 by the Association Technique Belge de Fonderie in collaboration with the Liége Section of the Association des Ingénieurs Sortis de l'Ecole de Liége. Abstracts of the papers on roll manufacture are given below; other abstracts will be found under "Rolling Mill Practice."—R. F. F.

History of the Manufacture of Rolls in Belgium. G. Nève. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 34–38). The author reviews the developments of the roll-making industry in Belgium, and shows the importance of this special branch of foundry work. After examining briefly the economics of the industry, the author enumerates the various types of rolls now manufactured.—R. F. F.

Manufacture and Utilization of Rolls. M. Borgerhoff. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 39–59). The manufacture and utilization of rolls are reviewed.—R. F. F.

Future Prospects of the Manufacture of Cast Iron Rolls. C. Halbart. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 60–65). The author examines, mainly from the point of view of the intrinsic qualities of metals, various types of roll, and with each type compares modern rolls with older types. He mentions the possibility of improvement by the addition of special elements and by heat-treatment. —R. F. F.

The Manufacture of a Good Quality Roll. Choice of Raw Materials and Melting Furnaces. G. Nève. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 89–95). The importance of the melting bath and the choice of a melting furnace for the manufacture of rolls are discussed.—R. F. F.

The Manufacture of Hardened Cast Iron Rolls. M. Borgerhoff. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 95–112). The author explains how to regulate the depth and degree of hardening when casting rolls and shows that the study of casting and thermal stresses should lead to more precautions being taken against scaling of rolls operating under water sprays. He gives an example of the results obtained by the use of high percentages of nickel and chromium in the manufacture of hard rolls for cold-rolling.—R. F. F.

Moulding and Casting Rolls. J. Varlet. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 112–118). The author describes the moulding and casting of rolls for sections, merchant bars, sheets, and billets.—R. F. F.

The Manufacture of Cast Steel Rolls at the Ougrée-Marihaye Works. R. Iweins. (Revue Universelle des Mines, 1945, series 9, vol. 1, Sept. 15, pp. 122–128). The author compares cast with forged rolls and shows the evolution of the present process of manufacturing cast rolls at the Ougrée-Marihaye foundries. Details of the various qualities of steels used are also given.—R. F. F.

Casting Steel Mill Rolls. (Steel, 1947, vol. 121, Sept. 22, pp. 80–82, 112). A series of photographs illustrates the process of casting steel rolls for mills at the Pittsburgh plant of the Mackintosh-Hemphill Co.—C. O.

Preventing Shrinkage of Gray Iron Castings. W. P. Sullivan. (American Foundryman, 1947, vol. 12, Aug., pp. 26–28). Practical tests which showed the value of tellurium corewashes as chill inducers for preventing porosity in grey-iron castings are described.—C. O.

Some Notes on Feeding. S. L. Finch. (Institute of British Foundrymen, June, 1947, Paper No. 877: Foundry Trade Journal, 1947, vol. 82, July 31, pp. 297–303; Aug. 7, pp. 319–324). An outline of the mechanism of crystallization, solidification, and subsequent shrinkage of metals is given and the factors affecting the flow of metal from feeder head to casting are discussed.—R. A. R.

The Influence of Production Flow on Moulding Methods in Iron Foundries and Its Effect on P.M.H. and General Efficiency. R. C. Shepherd. (Institute of British Foundrymen, June, 1947, Paper No. 883). Moulding work in general is broadly reviewed and related to appropriate production flow methods. Typical continuous and bank systems of moulding operation are described. The practice of segregating moulding operations for all classes of work is suggested as a means of improving production per man-hour.—R. A. R.

Blowing Cores on a Mass Production Basis. M. F. Degley. (Foundry, 1947, vol 75, Aug., pp. 72–74, 247). A short description, with illustrations, is given of the core shop of a foundry producing 450 tons of engine castings daily.—C. O.

Electronic Core Baking. R. W. Crannell. (Foundry, 1947, vol. 75, July, pp. 66–69, 240–241). The theory, technique, and applications of the dielectric baking of sand cores made with urea-formaldehyde binders are discussed.—C. O.

Some Notes on the Surface Drying of Moulds. A. Cracknell and F. Cousans. (Institute of British Foundrymen, June, 1947, Paper No. 884: Foundry Trade Journal, 1947, vol. 82, Aug. 7, pp. 313–315; Aug. 14, pp. 343–347; Aug. 21, pp. 365–369). Investigations are described in which the effects of varying the volume and temperature of the air passing over the surface of rammed sand moulds made in a standard box 18 × 18 × 6 in. deep were studied. For moulds dried at about 215° C. there appeared to be an optimum rate of air flow (180 cu. ft./min./sq. ft. of mould surface). which gave the maximum rate of drying.—R. A. R.

Conditions Affect Choice of Molding Method. P. Dwyer. (Foundry, 1947, vol. 75, Aug., pp. 82–83, 222–226). Points to be considered in choosing the method of making a mould for a particular casting are discussed briefly.—C. O.

The Casting of a Gray Iron Diesel Engine Liner. N. A. Kahn and B. N. Ames. (Foundry, 1947, vol. 75, Aug., pp. 66–71, 218–220). An illustrated description is given of a technique evolved for the mass production in a jobbing shop of a small grey-iron diesel-engine cylinder liner.—C. O.

Direct Rolling and Continuous Casting. Z. Wusatowski. (Hutnik, 1946, vol. 13, Feb., pp. 91–96; Mar., pp. 165–171). [In Polish]. The various methods used since Bessemer's original work on direct rolling are discussed. References are given to existing patents and technical papers.—W. J. W.

Centrifugal Casting. J. W. Moore and J. W. MacKay. (Mechanical Engineering, 1947, vol. 69, July, pp. 551–558). Descriptions are given of the techniques for the centrifugal casting of tubes in sizes up to 35 in. in outside dia. with walls up to 5 in. thick, in lengths up to 16 ft., of carbon and stainless steels. Ships' propeller shafts were successfully cast in this way. Details of the properties and microstructures of the steels are presented. —R. A. R.

Ferrous Die Castings. C. D. Pollard, H. A. Redshaw, and C. A. Payne. (Institute of British Foundrymen, June, 1947, Paper No. 874). Detailed descriptions are

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given of the Eaton Erb die-casting machine, the mode of operation, the making of dies, plant layout and organization, and the microstructure and properties of the iron castings produced.—R. A. R.

Producing High Purity Metals with Vacuum. J. D. Nisbet. (Iron Age, 1947, vol. 159, June 19, pp. 56–59). The design of furnaces for vacuum melting is discussed. A system incorporating an arrangement for loading and making additions while maintaining a vacuum of 0·00001–0·0001 in./Hg, a centrifugal casting apparatus (also maintained under the vacuum), and a 50-kW., 10,000-cycle induction furnace capable of melting 6 lb. of metal, is described and illustrated.—C. O.

Wetherill Vacuum Casting Process. W. A. Phair. (Iron Age, 1947, vol. 160, Aug. 14, pp. 67–70). A short description is given of the Wetherill-Armour vacuum casting process. The mould, arranged for bottom feeding, is placed inside a vacuum bell ; the bottom of the mould is connected by a tube to a ladle containing the molten metal, which is sucked into the mould cavity when the vacuum is formed. Advantages claimed for the process are : (1) Increased yield of good castings because of exact control over pouring conditions ; (2) thinner sections may be poured than by other methods ; (3) some improvement in yield because of less loss in runners ; and (4) a reduction in porosity and slag inclusions.—C. O.

Principles of Precision Investment Casting. K. Geist and L. M. Kerr, jun. (Institute of British Foundrymen : 1947, June, Paper No. 882). A detailed account is given of the principles and practice employed in the production of castings by the precision investment (or lost wax) process.—R. A. R.

Solidification Phenomena in Metals. W. F. Brandsma. (Metalen, 1947, vol. 1, Aug., pp. 231–234). [In Dutch].

Castings Should Always Be Qualified in the Foundry. O. O. Gammon. (Transactions of the American Foundrymen's Association, 1946, vol. 53, pp. 342–344). The addition of lugs and bosses to castings to facilitate setting-up in subsequent machining operations is advocated, and some examples are described.—R. A. R.

Foundry Grinding Wheels. C. A. Carlson. (American Foundryman, 1947, vol. 11, June, pp. 22–27). The manufacture, selection, and use of grinding wheels for foundry work are discussed.—C. O.

The Foundry Atmosphere. W. N. Witheridge. (Iron Age, 1947, vol. 160, Aug. 21, pp. 62–67 ; Aug. 28, pp. 67–73). Popular misconceptions regarding the ventilation of foundries are discussed. In the opinion of the author the delay in the adoption of available dust-control systems is due to a lack of understanding of the nature of specific problems. Several of the latter are considered in detail, and practical suggestions are made.—C. O.

Pricing Castings for Profit. J. B. Meier. (Foundry, 1947, vol. 75, Feb., pp. 72–73, 178–182 ; Mar., pp. 108–109, 243–248). The good and bad features of different methods of costing in the foundry are discussed.—R. A. R.

Foundry Costs and Cost Controls. C. E. Westover. (American Foundryman, 1947, vol. 11, July, pp. 44–48). The determination of the costs of production in foundries is discussed. Methods of calculating the loss due to reduced production in a plant operating at a rate less than its economical capacity and of determining rates for piece work are outlined.—C. O.

A Foundry Training Scheme. (Engineer, 1947, vol. 184, July 18, p. 65). The foundry-training centre recently opened by the Stanton Ironworks Company, Ltd., Nottingham, is briefly described.—C. O.

HEATING FURNACES AND SOAKING PITS

Fuel Oil in the Steel Industry. T. C. Bailey. (Institute of Petroleum Review, 1947, vol. 1, Aug., pp. 269–275). Applications of oil-firing for reheating and sheet annealing furnaces, and in the terne-coating process are described.—C. O.

Rolling-Mill Furnaces. G. Grenier. (Revue de l'Industrie Minérale, Comptes Rendus, 1944, July, pp. 67–71 ; Aug., pp. 91–96 ; Sept., pp. 104–108 ; Oct., pp. 111–116.) A review of reheating furnaces for rolling mills is presented.—R. A. R.

Electrometallurgy in Canada. J. L. Balleny. (Electrical Engineering, 1947, vol. 66, Aug., pp. 774–778). A general account is given of the ways in which the Canadian metallurgical industries utilize the electricity which is available at low cost from the hydro-electric plants. They include steel-strip annealing, wire patenting, production and heat-treatment of aluminium, induction heating, and melting steel in electric furnaces.—C. O.

Heat Losses in Furnace Linings. J. D. McCullough. (American Gas Journal, 1947, vol. 167, July, pp. 29–32 ; Aug., pp. 36–38 : Blast Furnace and Steel Plant, 1947, vol. 35, July, pp. 817–823). Methods for rapidly determining the heat lost by storage in or conduction from the linings of furnaces are outlined.—C. O.

Planning the Forge Shop for High-Frequency Heating. G. F. Applegate. (Steel, 1947, vol. 121, July 14, pp. 84–86, 124–128). The advantages of induction heating for forging operations are enumerated and the selection and maintenance of coils and high-frequency motor-generators are discussed.—C. O.

Metallurgical Uses of High-Frequency Furnaces in Sweden. L. Högel. (Fuel Economy Conference, World Power Conference, The Hague, 1947, Section C2, Paper No. 14). A short account is given of the construction of the high-frequency melting and heat-treatment furnaces employed by the Swedish metallurgical industries. Characteristic features of these furnaces are the use of series capacitors, automatic control, ceramic insulation of the coils, double-frequency coils, and the arrangement around the coils of packets of laminated transformer iron.—C. O.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Influence of Water Vapour and Methane on the Heat Treatment of Steel in a CO : H_2 Gas Atmosphere. H. M. Heyn. (Western Metal Congress : Metal Progress, 1947, vol. 52, Aug., pp. 232–237). A very brief summary is made of the subject of controlled heat-treatment atmospheres. Special mention is made of the control of water vapour and methane in gas mixtures containing approximately 20% of carbon monoxide, 40% of hydrogen, and 40% of nitrogen.—C. O.

Suspended Carburization. O. E. Cullen. (Steel Processing, 1947, vol. 33, July, pp. 431–435). A general description is given of a technique whereby the process of gas-carburization in a continuous furnace may be temporarily stopped and later continued without any harmful effect. The process involves maintaining the various zones of the furnace at different temperatures and very strict control of the gas composition.—C. O.

A Carburizing Experiment with Radioactive Carbon. J. K. Stanley. (Metal Progress, 1947, vol. 52, Aug., pp. 227–229). A short description is given of experiments in which the mechanism of pack carburization was studied by means of adding a small quantity of radioactive barium carbonate (containing the carbon isotope

of atomic weight 14) to the carburizing mixture. After the heating cycle about 85% of the radioactivity was found to have "disappeared," presumably by diffusion into the iron. The evidence is considered to support the theory of carbonate catalysis advanced by T. C. Fong and R. A. Ragatz.—C. O.

You Can Profit from Flame Hardening. (Australian Welding Institute : Industry and Welding : Australasian Engineer, 1947, July 7, pp. 77–78). Applications of flame-hardening are briefly summarized.—C. O.

Practical Aspects of Surface Hardening Methods. H. E. Boyer. (Iron Age, 1947, vol. 160, Aug. 14, pp. 74–79 ; Aug. 28, pp. 74–78). The practical advantages and limitations of pack carburizing, liquid carburizing, gas carburizing, and nitriding are compared. It is shown that, although there is some overlapping in the work which each process can economically perform, in general they are not competitive.—C. O.

Theory and Practice in Austempering. Part 1—Theory of Austempering. J. G. McMaster. (Australian Institute of Metals : Australasian Engineer, 1947, July 7, pp. 45–48). The alterations in crystal structure resulting from quenching, tempering, and austempering are illustrated.—C. O.

Theory and Practice of Austempering. Part 2—Practice in Austempering. J. N. Ellis. (Australian Institute of Metals: Australasian Engineer, 1947, July 7, pp. 49–54). The practical applications and advantages of austempering are outlined.—C. O.

Practical Aspects of Bainitic Hardening of High-Speed Steel. C. K. Baer and A. E. Nehrenberg. (Iron Age, 1947, vol. 160, Sept. 4, pp. 65–70 ; Sept. 18, pp. 77–82). The process of hardening high-speed steels known as "bainitic" hardening is briefly explained and compared with conventional practice. The changes in structure and physical properties of steels during the process are discussed. Data obtained in laboratory and service cutting tests are presented to show the improvements in tool performance resulting from bainitic hardening, and the influence on cutting efficiencies of primary and secondary bainite.—C. O.

Carbo-Nitriding of S.A.E. Steel Parts. W. H. Holcroft. (Metal Progress, 1947, vol. 52, Sept., pp. 380–386). An account is given of the process of carbo-nitriding, in which steels are case-hardened by the simultaneous absorption of carbon and nitrogen on heating in a suitable gaseous atmosphere (made by adding ammonia and a hydrocarbon gas to a mixture of 20% of CO, 40% of H_2, and 40% of N_2), and cooling. Typical results are illustrated by microphotographs.—C. O.

One Salt Bath for Both Carburizing and Cyaniding. F. Steigerwald. (Materials and Methods, 1947, vol. 26, July, pp. 75–77). A description is given of a new type of salt bath which can be used for both carburizing and cyaniding, and permits close control of case depth, rapid carburization, and easy cleaning of parts after quenching.—J. R.

Industrial High-Frequency Heating. J. C. van Reysschoot. (Revue Universelle des Mines, 1946, Series 9, vol. 2, No. 7, pp. 273–278). High-frequency heating, for the case-hardening and welding of metals is discussed.—R. F. F.

Deformation of Machine Steel Rings during Case Hardening. J. E. Erb. (Steel Processing, 1947, vol. 33, June, pp. 347–349, 368–369). The data obtained during a study of the change in dimensions of rings (6-in. outside, 2-in. inside, dia.) of low-carbon steel during case-hardening are tabulated. It is shown that the bores tended to assume an "hour glass" shape and that the outer surfaces tended to become barrel-shaped. The effect of previous stress-relieving was found to be small.—C. O.

The Art of Quenching ; 1—Coolants. R. G. Roshong. (Metal Progress, 1946, vol. 50, Nov., pp. 948–951). After a short historical survey, the characteristics and applications of each of the following quenching media are discussed: Water, brine, oils, molten salts, and molten lead.—C. O.

FORGING, STAMPING, DRAWING, AND PRESSING

Cold-Formed Steel Wire Springs. (Wire Industry, 1947, vol. 14, Aug., p. 447). A very short account is given of the production of steel wire springs for the upholstery and similar industries.—C. O.

The Production of British Clearing Heavy Presses. (Machinery, vol. 71, Aug. 7, pp. 143–149). The article describes the production of heavy presses (with plant adapted from its wartime use in the manufacture of naval gun mountings), of capacities up to 1500 tons and weighing up to 350 tons, by Vickers Armstrongs, Ltd. The press components are fabricated from steel plates and slabs by electric welding. The heat-treatment and the tools for machining the components are also described.—R. L. B.

Cemented Carbide Used for High Production Dies and Punches. (Iron Age, 1947, vol. 160, Aug. 28, pp. 64–66). Examples of increased tool life resulting from the use of cemented carbides (in particular, Carboloy) instead of steel for punches and dies are cited.—C. O.

Forging Die Design—The Die Block and Preparation for Use. J. Mueller. (Steel Processing, 1947, vol. 33, July, pp. 424–426, 430, 436). The manufacture and mounting of alloy-steel forging-die blocks are discussed.—C. O.

Wright Wire Company, The Germ of Wickwire-Spencer. K. B. Lewis. (Wire and Wire Products, 1947, vol. 22, Aug., pp. 588–590, 614–619). The history of the Wright Wire Co., Worcester, Mass., is outlined.—C. O.

Borax in Steel Wire-Drawing Operations. (Borax Consolidated Limited, Technical Service Bulletin, 1947, Sept. 22, No. 9). The use of a hot 5–7% borax solution as an alternative to lime is advocated for wire drawing, and the technical advantages are listed. The preparation of the borax bath and the control of borax solutions are described.—M. A. V.

Use of Wire Drawing Compounds. E. L. H. Bastian. (Wire and Wire Products, 1947, vol. 22, Aug., pp. 577–580). A survey is made of the duties and characteristics of the common wire-drawing compounds.—C. O.

Surface Defects on Bars for Bolts—Manufacture and Drawing. R. Hoffmann. (Revue Universelle des Mines, 1946, Series 9, vol. 2, No. 9, pp. 405–411). The defects occuring on bars used for heading and drawing, and methods of prevention are described.—R. F. F.

Quality Wire—Some Factors in Its Achievement. (Wire Industry, 1947, vol. 14, Sept., pp. 497–501). The influence of plant and process control and personnel management on the quality of the wire produced is discussed.—C. O.

Design Considerations for Carbide Sheet Metal Dies. E. Glen. (Steel, 1947, vol. 121, Sept. 22, pp. 75–76, 109). A short summary is made of factors affecting the design and fabrication of carbide dies and punches for drawing sheet metals.—C. O.

Diamond Dies for Wire Drawing. Sir C. Paterson and R. E. Leeds. (Research, 1947, vol. 1, Oct., pp. 2–10). An illustrated description is given of the mechanical methods used for piercing diamonds for wire-drawing dies.—C. O.

ROLLING-MILL PRACTICE

Electronic Control, Using Rectifiers, for the Non-Reversing Rolling Mill at the Pompey Steelworks. (Génie Civil, 1947, vol. 124, Sept. 15, pp. 351–353). This rolling-mill is driven by a 750-V. D.C. motor and mercury-vapour rectifiers are used to convert the three-phase A.C. to D.C. These rectifiers are described.—R. F. F.

Life, Defects, and Scrapping of Rolls—General. H. Bersoux. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 65–71). This article makes a general examination of the causes of the short life of rolls, with reference to the necks, body, and wobbler. The use of synthetic-resin bearings is discussed both from the technical and economic point of view. The wear and replacement rates are given for various types of mills with notes on maintenance.—R. F. F.

Life, Defects, and Scrapping of Rolls—Rolls for Sections. E. Renaud. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 72–79). The author refers to the various qualities of rolls and indicates the best uses for them. He gives details of improvements to be made in rolling mills and roll maintenance, and comments on the more common causes of fracture in rolls.—R. F. F.

Life, Defects, and Scrapping of Sheet Mill Rolls Working under Water Sprays. C. Colbrant. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 79–85). The major defect of rolls working under water sprays is scaling. Statistical study shows that the depth of hardening is of great importance.—R. F. F.

Some Considerations of the Nature and Efficiency of Rolls for Billet Mills and Continuous Billet-, Wire- and Strip-Mills. A. Lesgardeur. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 85–88). The choice of the quality of rolls depends on local conditions, such as the type of mill and the maintenance routine. Examples are given together with mill output data.—R. F. F.

Care of Hot Sheet Mill Rolls. J. Varlet. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Aug. 15, pp. 118–121). The author indicates precautions which should be taken to prevent the premature fracture of quenched cast-iron rolls, and emphasizes the importance of the direction of rotation of rolls which should not be left to chance, and gives details of the maintenance needed before, during, and after rolling, with special emphasis on the cooling period of rolls.—R. F. F.

Rolls for Cold-Rolling. P. Mélon. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Sept. 15, pp. 128–137). Details are given of metals which can be cold-rolled, together with the uses, quality, and hardness of cold-rolled strip steels. The composition, hardness, fatigue, maintenance, and defects of steel rolls for cold-rolling are described.—R. F. F.

Elementary Analysis of Rolling Theory. G. Moressée, (Revue Universelle des Mines, 1945, Series 9, vol. 1. Sept. 15, pp. 137–146). The ductility of metal is defined; the movement of a metallic particle during a pass is analysed; the theoretical energy requirements are computed, and the stresses to which the rolls are subjected are discussed.—R. F. F.

The Dimensions of Steel Springs. A. Schillak and F. Gerlich. (Hutnik, 1946, vol. 13, Feb., pp. 82–85). [In Polish]. Details of roll calibration are given with special reference to the Gerlich method.—W. J. W.

Modernized Control System Increases Blooming Mill Output. E. Frisch. (Steel, 1947, vol. 121, Sept. 15, pp. 120–125). Details are given of a rotating regulator control for a 44-in. blooming mill which has enabled production to be increased by 20% because of the reduction in the time required for reversing the mill as compared with the old reverse-plugging system.—C. O.

Pressure Impulses in Pipe Lines. O. Schnyder. (Von Roll Mitteilungen, 1943, vol. 2, Feb.–June, pp. 1–56). The author deals comprehensively with the problem of pressure impulses in pipelines. The paper is in three parts, namely: (1) The principles and theory of pressure impulses; (2) practical impulse problems; and (3) pressure impulses in non-cylindrical pipelines with high rates of flow.—R. A. R.

Cogging. M. Holzweiler. (Stahl und Eisen, 1947, vol. 66–67, Sept. 11, pp. 315–318). The calibration of cogging-mill rolls, and the size and direction of ribs on the roll face to grip the ingot are discussed with examples of old and recent calibrations.—R. A. R.

Rolling Mills. (Combined Intelligence Objectives Sub-Committee, 1947, File No. XXXIII-70: H.M. Stationery Office). This report, which forms Appendix 19 to C.I.O.S. Report XXXII-119, contains short descriptions of special-section rolling mills at the works of Bochumer Verein, Weitmar, of a hot Steckel mill at the Hoesch Works, Dortmund, of rolling mills at the Hermann Goering Works, Watenstedt, and of the Roeckner mill.—C. O.

The Development of Continuous Hot and Cold Rolling of Flat Rolled Iron and Steel Products. S. E. Graeff. (South Wales Institute of Engineers, Oct. 16, 1947). After a survey of the trend and development of the American steel industry the author traces the history of the continuous mill for the production of sheet and strip and points out a number of factors which determine the successful operation of continuous mills. The mill invented by J. B. Tytus and put down by the American Rolling Mill Company at their Ashland plant in 1923 produced sheets in multiple lengths up to 30 ft. with intermediate heating. The continuous mill put down in 1926 at Butler, Pennsylvania, which became the prototype of the present-day mill, used no intermediate heating and produced very long lengths which could be coiled or sheared into set lengths as desired. The continuous sheet and wide strip mills at present operated in the United States are given in a table and the development and operation of four-high mills and cluster mills such as the Sendzimir mill are reviewed. Attention is drawn to the importance of tension in cold rolling and particulars are included of the rolls used in the hot and cold mills at Ebbw Vale.—R. E.

Heating and Rolling Low Carbon Steel for Strip. N. E. Rothenthaler. (Iron and Steel Engineer, 1947, vol. 24, Aug., pp. 71–76). A short survey is made of modern practice in the hot-rolling of low-carbon steel strip.—C. O.

Tube Straightening. E. W. Wrage. (Mechanical Engineering, 1947, vol. 69, Aug., pp. 648–650). The principles of the press and roll methods of tube-straightening are illustrated.—C. O.

LUBRICANTS AND LUBRICATION

New Studies of the Lubrication of Surfaces at Very High Temperatures. P. Martinet. (Revue de l'Industrie Minérale, Mémoires, 1945, Mar., pp. 71–80). The lubrication of cylinders of engines driven by blast-furnace gas is discussed.—R. F. F.

Roll Oil Emulsion for Cold Strip Rolling. R. W. Piper. (Iron and Steel Engineer, 1947, vol. 24, July, pp. 62–63). As an example of the use of an oil emulsion as a roll lubricant, the lubrication system of a 42-in. reversing cold-strip mill is described. The optimum composition of the emulsion is considered to be 25% of roll oil and 75% of water, maintained at a *p*H value of 7·5–9.—C. O.

Surface Activity of Lubricants. J. M. Wilson. (Iron and Steel Engineer, 1947, vol. 24, Aug., pp. 46–49). The anti-frictional and corrosion-resisting properties of lubricating oils are discussed with reference to the elementary theory of surface activity.—C. O.

Centralized Lubrication. A. J. Jennings. (American Society of Lubrication Engineers: Iron and Steel, 1947, vol. 20, July, pp. 345–349; Sept., pp. 437–441). A short account is given of the history of centralized lubrication and a number of systems and devices used today are described.—C. O.

WELDING AND FLAME-CUTTING

Atomic Hydrogen Welding. (Aircraft Engineering, 1946, vol. 8, Nov., pp. 533–535). A brief account is given of practice in the use of atomic-hydrogen welding at the plant of the Ryan Aeronautical Co. in the United States, for the production of stainless-steel exhaust manifolds. A chart is also included showing the relative amounts of carbon absorbed by the parent metal in welding by various processes.—R. E.

Pressure Welding. E. R. Proctor. (Aircraft Production, 1947, vol. 9, Feb., pp. 63–66). The application of the Uniweld process for the welding of aircraft landing gear is described and illustrated. The advantages of the process for this purpose are pointed out and particulars are given of the welding equipment used.—R. E.

The Effect of Welding Technique on Brittle Transition Temperature. N. Grossman and P. R. Shepler. (Welding Journal, 1947, vol. 26, June, pp. 321-s–331-s). The experimental details of an investigation of the effects of 12 different welding techniques on the temperature of transition from tough to brittle fracture of welded high-tensile silicon steel plates are reported. The transition temperatures were determined by subjecting notched-bar specimens, supported horizontally on two knife edges, to central bending loads at decreasing temperatures, until the specimens failed with only a slight amount of yield. The results indicate that the techniques can in general be divided into two groups—those which bring about a loss of ductility in the heat-affected zone or the weld metal, and those in which the weld metal is more ductile and the heat-affected zone only slightly less ductile than the base metal. Variation in the speed of welding, within the limits of commercial practice, had little effect on ductility.—C. O.

Rational Production Methods for Steam Piping and Welded Pipe Lines. Parys. (Mémoires de la Société Royale Belge des Ingénieurs et des Industriels, 1945, Série B, No. 2, pp. 343–346). Recent progress in the production of welded steam pipe lines is reviewed.—R. A. R.

Steel Tube Manufacture and Pipe Fabrication. J. Atherton. (Australian Welding Institute: Australasian Engineer, 1947, June, 7. pp. 68–70). Methods used for butt-welding tubular sections are outlined.—C. O.

Welding and Weldability of Steels. G. Grenier. (Revue de l'Industrie Minérale, Comptes Rendus, 1946, Aug.-Sept., pp. 129–134). This is a review of current practice.—R. F. F.

Welding Processes in the Manufacture of Tools. A. Leroy. (Mémoires de la Société des Ingénieurs Civils de France, 1941, vol. 94, pp. 481–511). A review of the methods and materials in present-day use is presented.—J. C. R.

Saving of High-Speed Steel by the Use of Tools with Inserted Tips. J. Jacquet. (Mémoires de la Société des Ingénieurs Civils de France, 1941, vol. 94, Oct.–Dec., pp. 512–540). The author compares three methods of making high-speed steel tools: (*a*) With the whole head of high-speed steel; (*b*) with a blade of high-speed steel let into the head; and (*c*) with a tip of high-speed steel welded on the head.—J. C. R.

Welding in the Repair and Maintenance of Tools. R. Meslier. (Mécanique, 1946, vol. 30, May, pp. 125–127). The use of welding for surface-hardening and tipping tools is described.—R. F. F.

Repair Procedures in a Production Weld Shop. O. E. Brown and C. R. Causey. (Iron Age, 1947, vol. 159, June 19, pp. 60–63). A system for handling repair work in a welding shop without interfering with regular production is suggested. The repairs are classified according to composition (*i.e.*, mild steel, cast iron, low-alloy steel, and stainless steel), and a standard procedure is followed for each class.—C. O.

The Technique of Welding in Steel Construction. A. Dunoyer. (Mémoires de la Société des Ingénieurs Civils de France, 1945, vol. 98, Jan.–Mar., pp. 78–89). A brief review is given of progress in welded steel construction.—J. C. R.

Automatic Fusion Welding. (Electrical Review, 1947, vol. 141, Aug. 15, pp. 237–242). An example is given of the application of the very heavy-duty automatic fusion-welding equipment installed by Messrs. Babcock and Wilcox, Ltd. The installation is mainly used in the manufacture of steam drums for high-pressure boilers.—J. R.

New Full and Semi-Automatic Welding Machines. P. Sevbo and V. Paton. (Institute of Electro-Welding, Ukrainian Academy of Sciences, 1946, Booklet: Engineers' Digest, N.Y., 1947, vol. 4, July, pp. 314–322). Following a critical survey in 1942 of existing welding machinery in Russia, three standard types have been developed: (1) The combined welding automat UWA-2; (2) the "welding tractor" TS-6; and (3) the welding "semi-automat TS-8." These machines, of which the first two are now widely used, are described.—M. A. V.

Three-Phase Welding. J. L. Solomon. (American Welding Society: Steel, 1947, vol. 121, July 7, pp. 94–98, 137–140). The three-phase single-phase method of resistance welding is compared with other methods from the point of view of the uniformity of the load on the power line and the power demand.—C. O.

Some Fundamental Principles for the Resistance Welding of Sheet Metal. H. E. Dixon. (Sheet and Strip Metal Users' Technical Association: Sheet Metal Industries, 1947, vol. 24, Mar., pp. 607–617, 624; Apr., pp. 813–820; June, pp. 1221–1226, 1230; July, pp. 1430–1435). The work of previous investigators on the subject of pressure welding is reviewed. The conclusions reached are applied to the study of the fundamentals of resistance welding, with particular reference to the spot and flash-butt welding of sheet metal.—C. O.

A Preliminary Investigation of the Spot Welding of Scaly and Rusty Structural Steel. W. D. Doty and W. J. Childs. (Welding Journal, 1947, vol. 26, June, pp. 358-s–362-s). A report is made on a study of the spot-welding of steel coated with scale, and scale plus rust. Satisfactory welds were obtained in rust-free scale-covered steel when a preheating sequence was included in the welding cycle. The presence of rust resulted in greater variation in the quality of the welds and in more rapid electrode deterioration.—C. O.

Production of High Quality Flash Butt Welds. J. S. Blair. (Transactions of the Institute of Welding, 1947, vol. 10, Aug., pp. 107–117). An outline is given of techniques for making flash-butt welds of high quality. Testing and control instruments are also considered.—C. O.

Inert-Arc Welding with Direct Current. R. F. Wyer. (Iron Age, 1947, vol. 160, Aug. 21, pp. 68–70). Details are given of the technique for welding stainless steel, copper, aluminium, and magnesium by the inert-arc process, using direct current.—C. O.

Coordinated Development of A.-C. Welders and Electrodes. C. P. Croco. (Welding Journal, 1947, vol. 26, Aug., pp. 680–683). An outline is given of the development of A.C. welding apparatus in the United States. —C. O.

Note on the Influence of the Water-Content of an Electrode-Coating on the Hydrogen-Content of Weld Metal. W. P. van den Blink. (Welding Journal, 1947, vol. 26, July, pp. 369-s–370-s). It is shown that the addition of water to an arc atmosphere does not necessarily raise the partial pressure of hydrogen in that atmosphere.—C. O.

Progress Report on the Spot Welding of High-Tensile Carbon and Low-Alloy Steels. W. F. Hess, W. D. Doty, and W. J. Childs. (Welding Journal, 1946, vol. 25, Sept., pp. 503-s–512-s). An account is given of experiments made to determine the optimum welding and tempering conditions for the spot-welding of 0·031-in. and 0·018-in. sheets of high-tensile steels containing small amounts of nickel, chromium, and molybdenum. Tempering in the welding machine was found to cause a very marked improvement in the mechanical properties of the spot welds, particularly in the thinner gauge sheets. The second part of the report gives details of an investigation of the effect of cold-rolling and heat-treatment on the properties of spot welds in a low-alloy steel, NE 8630, in 0·062-in. sheets. The properties of welds in normalized material were better than those in material cold-worked or heat-treated to higher tensile strengths. Cold-rolling after annealing impaired the welding properties of the steel, but cold-rolling after normalizing was beneficial. Better spot welds were obtained in steel the strength of which had been raised by quenching and tempering than in steel strengthened by cold-rolling.—C. O.

Production Jobs for Inert Gas Shielded Arc Welding, H. T. Herbst. (American Welding Society : Steel, 1947, vol. 121, Sept. 1, pp. 72–74, 92–96). A short account is given of the general characteristics and applications of the Heliarc welding process.—C. O.

Welding and Marine Engineering—An Examination of Present Tendencies. A. C. Hardy. (Welding, 1947, vol. 15, Sept., pp. 399–407). The future of welding in the fabrication of marine engines is discussed in relation to recent trends in design.—C. O.

British Shipbuilding—2. The Harland and Wolff Yards, Belfast. E. Cuthbert and D. Rebbeck. (Welding, 1947, vol. 15, Sept., pp. 429–440). The rapid expansion of the electric-welding facilities in the Harland and Wolff Yards, Belfast, in 1940, and present layout and practice are described.—C. O.

A Welded Locomotive Boiler. P. V. Norlin. (Teknisk Tidskrift, 1947, vol. 77, Sept. 20, pp. 685–689). [In Swedish]. The design of a Swedish locomotive boiler is discussed with details of the welding specifications.—R. A. R.

Belgian Specifications for Steels for Welded Structural Work. R. A. Nihoul. (Ossature Métallique, 1947, vol. 12, Jan., pp. 35–48). Details are given of specifications issued by the Institut Belge de Normalisation for light and heavy plates, sheets, and rolled bars and sections. The welding conditions for these steels, approved by the Comité Mixte des Aciers, are presented and discussed.—J. C. R.

Correct Technique Necessary for Stainless Welding. L. K. Stringham. (Iron Age, 1947, vol. 160, Aug. 28, pp. 61–63). Details are given of the correct techniques for welding stainless steels of the following four types: (*a*) Hardenable ferritic ; (*b*) partially hardenable ferritic ; (*c*) non-hardenable ferritic ; and (*d*) austenitic.—C. O.

Joining Corrosion-Resistant Alloys by Submerged Melt Welding. R. J. Anderson and H. J. Roberts. (Materials and Methods, 1947, vol. 26, July, pp. 89–92 ; Welding Journal, 1947, vol. 26, Apr., pp. 338–342). Details are given of techniques which have proved successful for the submerged-melt welding of corrosion- and heat-resistant alloys, including nickel-clad steel. —C. O.

Pressure Welding of Stainless Steel. A. L. Rustay. (Metal Progress, 1947, vol. 52, Aug., pp. 238–242). Details are given of the technique for the solid-phase welding of stainless steel.—C. O.

The Electric Welding of Pipe. O. Blodgett. (Welding Journal, 1947, vol. 26, Aug., pp. 684–689). The various systems of pipes fitted in ships are classified, and the electric-welding techniques used for joining each type are briefly indicated.—C. O.

Electric Resistance Welded Steel Tubing. E. W. Allardt. (Iron and Steel Engineer, 1947, vol. 24, Aug., pp. 55–64). The equipment and techniques used at the Alliance Ohio, plant of the Babcock and Wilcox Tube Co. for the production of resistance-welded steel tubes are described. —C. O.

Results Obtained with Radiography in the Detection of Defects in Welds. H. Louis. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 163–171). Types of defects in welds shown by X-rays are described and radiographic inspection is correlated with fatigue testing.—R. F. T.

Spot Welding Steel with an Alloy Content. C. A. Kershaw. (Welding, 1947, vol. 15, Aug., pp. 350–355). A short account is given of an investigation into the cause of failure of spot welds in low-carbon steel containing nickel, copper, and chromium as residuals. The trouble was found to result from local hardening, which could be removed by introducing a short post-heat into the welding cycle.—C. O.

Quality Control and Inspection of Welds. G. C. Close. (Steel Processing, 1947, vol. 33, June, pp. 350–354). The factors of design, welding technique, and inspection which enter into the control of the quality of welding are discussed.—C. O.

Metal Joining and Induction Heat Treating in Technical Metal Processing, Inc., Plant. (Industrial Heating, 1947, vol. 14, July, pp. 1006–1012, 1046–1048). The brazing, soldering, and carburizing techniques and equipment of Technical Processing, Incorporated, Cleveland, Ohio, are described.—C. O.

Brazing in Electric Furnaces. S. De Domenico and J. A. Comstock. (Metal Progress, 1947, vol. 52, July, pp. 71–76). The selection of atmospheres for brazing in electric furnaces is discussed. Recommendations are made for the atmospheres and types of brazing alloy to be used when brazing S.A.E. steels, high-alloy steels, nickel alloys, and copper alloys.—C. O.

Furnace Brazing 12 Per Cent. Chromium Low Carbon Steel. T. H. Gray. (Steel, 1947, vol. 121, July 21, pp. 104–106, 124–127). Tests to determine the effect of brazing conditions on the strength of brazed joints in 12% chromium low-carbon steel are described. It is concluded that the factors which contribute to the production of joints of the highest strength are : (1) Minimum spacing ; (2) capillary feeding ; (3) use of oxygen-free copper or copper-base alloys as brazing metals ; (4) avoidance of sand-blasted or other contaminated surfaces ; and (5) the use of a very pure cracked-ammonia atmosphere in the furnace.—C. O.

A Survey of Modern Welding Equipment. C. G. Bainbridge. (Iron and Coal Trades Review, 1947, vol. 155, Aug. 22, pp. 353–358).

Progress in the Metallurgical Industry in the U.S.A. A. Mercier. (Revue Technique Luxembourgeoise, 1947, vol. 39, Jan.–Mar., pp. 2–15). A critical review of modern American methods of fabrication is presented. —R. F. F.

Safe-Ending Large Diameter Locomotive Flues. (Steel, 1947, vol 121, July 28, p. 75). A flash-welding machine capable of butt-welding ends on to 5½-in. dia., locomotive boiler flues at the rate of 60 per hr., is briefly described. It is used in combination with a "stripper," which removes the material upset during the welding operation. —C. O.

Regulations for the Welding of Boilers and Steam Storage Vessels. P. Bongers. (Nederlandse Vereniging voor Lastechniek, Lassymposium, 1947, June, pp. 5–28). [In Dutch].—R. A. R.

The Welding of Boilers and Pressure Vessels. L. Capel. (Nederlandse Vereniging voor Lastechniek, Lassymposium, 1947, June, pp. 29–50). [In Dutch]. Recommendations on the welding technique for pressure vessels, including high-pressure steam pipes and fittings are made.—R. A. R.

The Welding of Unfired Pressure Vessels. W. P. Kerkhof. (Nederlandse Vereniging voor Lastechniek, Lassymposium, 1947, June, pp. 79–105). [In Dutch]. "The A.P.I.-A.S.M.E. Code" (American Petroleum Institute and American Society of Mechanical Engineers) for the welding of unfired pressure vessels is critically reviewed.—R. A. R.

Improved Welding Techniques Spark Success of Steel Plate Fabricator. (Steel Processing, 1947, vol. 33, June, pp. 358–359). A short description is given of the steel-fabrication plant of Black, Sivalls, and Bryson, Inc., Oklahoma City, Oklahoma. More than 130 Lincoln "Shield Arc" welding machines, 6 automatic welding machines, and welding "positioners" of 25,000 lb. capacity are used in the fabrication of pressure vessels up to 60 ft. in length and 10 ft. in dia.—C. O.

Flame Cutting Structural Shapes in Quantity. (Steel, 1947, vol. 121, July 21, pp. 100–101). An apparatus for flame-cutting steel joists and channels in large quantities at speeds of 4–24 in./min. is illustrated.—C. O.

Electro-Erosion Cutting of Hard Metal and Steel. T. P. Rekshinskaya. (Automobilnaya Promyshlennosti, 1946, No. 5-6, pp. 12–15: Engineers' Digest, N.Y., 1947, vol. 4, July, pp. 304–305). An experimental investigation into the cutting of hard metals by electro-erosion (*i.e.*, the destruction of a metal surface by means of an electric spark discharge) has been concluded at the Gorki Automobile Works. The machine is described and details of its performance given with some comparative figures for sawing. These show that electro-cutting may be used for hard metal where sawing is quite impracticable.—M. A. V.

Oxyarc Cutting—An Important New Development. R. N. Thompson. (Welding, 1947, vol. 15, Aug., pp. 344–349). An account is given of the "oxyarc" cutting process and its applications. The process consists of striking an arc between the part to be cut and a hollow steel electrode; when the metal is heated sufficiently oxygen at high pressure is directed on to it through the electrode.—C. O.

Metal Cutting Techniques. O. W. Boston. (Iron Age, 1947, vol. 160, Sept. 11, pp. 185–189). A general survey is made of present-day cutting tools and techniques in relation to possible future requirements.—C. O.

Mechanisms of Metal Cutting. W. H. Oldacre and H. A. Erickson. (Mechanical Engineering, 1947, vol. 69, Aug., pp. 655–657). The character and structure of the chips formed during the cutting of steel with a single-edged tool have been investigated. The appearance and microstructures of the chips formed at six different cutting speeds are illustrated.—C. O.

MACHINING AND MACHINABILITY

Brinell Hardness *versus* Machining Index as a Criterion of Machinability. G. Schlesinger. (Machinery, 1947, vol. 71, July 31, pp. 122–127). The author illustrates his contention that the Brinell hardness of a metal is not a criterion of machinability by references to data in the American literature. The best indication of the cutting resistance of a material is considered to be the machinability index, T, which is the tangential force required for the removal of chips, expressed in lb./0·001 sq. in. Numerical examples are given of the application of the machinability index to such calculations as power consumption and rate of chip removal. —C. O.

The Machinability of Steel. (Mechanical World, 1947, vol. 122, Aug. 8, pp. 131–133). Tests for determining machinability are discussed, in particular an adaptation of the "Meyer Analysis" in which the rate of penetration of the ball indenter in relation to the load is observed. Progress in the improvement of machinability is expected to be in the direction of introducing sub-microscopic dispersions of non-alloying inclusions.—R. A. R.

Effect of Microstructure on Machinability of Cast Irons. M. Field and E. E. Stansbury. (Transactions of the A.S.M.E., 1947, vol. 69, Aug., pp. 665–682). Quantitative determinations have been made of the relative machinabilities of cast irons with various microstructures, in terms of tool life, power requirements, and surface finish.—C. O.

Modern Developments in Precision Boring Practice. C. A. Alden. (Iron Age, 1947, vol. 160, July 31, pp. 50–55). The adaptability and economy of the precision-boring process is illustrated by several practical examples of single and multiple boring operations.—C. O.

Developments in the Field of High Speed Steels. J. Cournot. (Mémoires de la Société des Ingénieurs Civils de France, 1941, vol. 94, Oct.–Dec., pp. 471–480). The author traces the development of molybdenum high-speed steels, examines their properties and compares the results of cutting tests with those of steels having a higher tungsten content.—J. C. R.

How to Use Carbide Cutters for Milling. H. A. Frommelt. (Iron Age, 1947, vol. 159, Feb. 13, pp. 44–49; Feb. 20, pp. 48–53; Feb. 27, pp. 46–52; Mar. 6, pp. 74–78; Mar. 13, pp. 52–56; Mar. 20, pp. 56–59; Mar. 27, pp. 58–62; Apr. 10, pp. 78–81; Apr. 17, pp. 58–63; Apr. 24, pp. 58–62; May 8, pp. 75–79; May 15, pp. 62–65; May 22, pp. 77–79; June 5, pp. 80–83; June 12, pp. 65–68; June 19, pp. 77–80; vol. 160, July 3, pp. 69–72; July 10, pp. 66–68; July 17, pp. 61–64; July 24, pp. 68–70). A comprehensive survey is made of steel-milling practice with sintered carbide cutters.—C. G.

Manufacture of Balls for Bearings and Pens. J. Porterfield. (Materials and Methods, 1947, vol. 25, June, pp. 97–100). A general account is given of the processes involved in the manufacture of high-precision steel balls—pressing, removal of flash, rough-grinding, multiple-groove grinding, tumbling, heat-treatment, precision grinding, lapping, inspection, and gauging.—C. O.

Wet Grinding Sheet and Strip. T. E. Lloyd. (Iron Age, 1947, vol. 160, Aug. 14, pp. 65–66). A short account is given of a machine for simultaneously wet-grinding both sides of cut or coiled sheet and strip. The grinding

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is performed by a diagonal oscillatory movement of abrasive bands wound on pneumatic rubber drums.—C. O.

Finishing with Diamond Tools. P. Grodzinski. (Mechanical World, 1947, vol. 121, June 27, pp. 615–616). Machining with single-point cutting tools with a shaped diamond is shown to have many advantages over older methods, particularly for accurate and highly finished work. The grinding and polishing of hard materials by means of bonded or loose diamond abrasive is discussed—G. C. J.

Use of Soluble Oil Coolants. L. B. Johnson. (Steel, 1947, vol. 121, July 7, pp. 110–113, 142). Data are given on the compositions of oil emulsions for use as cooling fluids in machining operations on ferrous and non-ferrous metals.—C. O.

CLEANING AND PICKLING

Measurement of Metal Polish Performance. F. E. Clarke and R. C. Adams. (A.S.T.M. Bulletin, 1947, Aug., pp. 57–62). An account is given of the design and use of a machine for comparing the performance of various metal polishes.—C. O.

Industrial Developments in the Electrolytic Polishing of Metals and Alloys. P. A. Jacquet. (Métaux et Corrosion, 1940, vol. 15, Sept.–Oct., pp. 68–77). The history of electrolytic polishing is reviewed and the processes in use for the polishing of nickel, iron, steels, and stainless steels are outlined. The corrosion resistance of electrolytically polished surfaces is discussed and possible future industrial applications are considered.—J. C. R.

Preparation of Metals for Painting—A Review. R. E. Gwyther. (Modern Metal Protection Symposium: Corrosion, 1947, vol. 3, Apr., pp. 201–207). A summary is made of the physical and chemical methods used for cleaning metals before painting.—C. O.

Treatments of Metal Surfaces Prior to Painting. E. F. Hickson and W. C. Porter. (Product Engineering, 1947, vol. 18, Aug., pp. 128–130). Recommended methods for the cleaning of steel and various non-ferrous metals before painting are outlined.—C. O.

PROTECTIVE COATINGS

Wear-Resistance of Electrodeposited Chromium Coatings. J. Desmurs. (Métaux et Corrosion, 1940, vol. 15, Mar.–Apr., pp. 32–34). A brief account is given of tests carried out by various investigators.—J. C. R.

Chevrolet Increases Nickel Plating on Bumpers. L. Strong and H. F. Reves. (Products Finishing, 1947, vol. 11, Aug., pp. 20–26). A description is given of the electroplating plant of the Chevrolet Division of the General Motors Corporation, Detroit, in which a nickel coating 0·0015 in. thick is deposited on steel motor-car bumpers.—C. O.

The Protection of Ferrous Metals by Galvanizing. J. Frasch. (Métaux et Corrosion, 1940, vol. 15, May–June, pp. 37–42; July–Aug., pp. 51–56). After outlining the principles of galvanizing and its field of application, the author considers in detail the practical processes involved, with particular reference to galvanizing ferrous metals.—J. C. R.

Continuous Galvanizing. A. D. Stout, jun. (Iron Age, 1947, vol. 160, Aug. 28, pp. 79–80). Features of a process developed by the National Tube Co., Lorain, Ohio, for the continuous hot-dip galvanizing of 2-in. pipe at the rate of 600 tons per day are described.—C. O.

Conditioning Hot Dip Galvanizing Baths. W. G. Imhoff. (Steel, 1947, vol. 121, Aug. 18, pp. 108, 125–127). The control of the conditions in zinc baths to ensure the satisfactory hot-dip galvanizing of small articles, without "sticking" or "freezing" is discussed.

Surface Texture Study of Electroplated Zinc. R. L. Lunt. (Metal Finishing, 1947, vol. 45, Aug., pp. 68–69). The results are presented of an investigation of zinc and zinc-alloy plating from acid baths; the paper presents a photomicrographic study of electroplated surface textures. Factors influencing surface texture and smoothness on strip and sheet steel are given.—G. C. J.

Prevention of Sulphur Staining. (Tin Printer and Box Maker, 1947, vol. 23, Sept., pp. 4–6). The principal methods used for preventing "sulphur blackening" on the inside of tin-plate food containers are outlined.—C. O.

Spra-Bonderizing. (Automobile Engineer, 1947, vol. 37, Aug., pp. 313–314). A short description is given of the continuous spraying process used by Wolseley Motors, Ltd., for the phosphate-coating of motor-car bodywork.—C. O.

Electrostatic Finishing Process Doubles Production at Delta Electric. W. B. Stephenson. (Products Finishing, 1947, vol. 11, Aug., pp. 40–46). The electrostatic process used by the Delta Electric Company, Marion, Indiana, for the enamel-coating of bicycle lamps is briefly described.—C. O.

A New Method of Control of the Porosity of Coatings of Paint. J. Frasch. (Métaux et Corrosion, 1940, vol. 15, Mar.–Apr., pp. 28–31). A simple and rapid method is described for detecting porosity in coatings of paint. A painted metal plate and a carbon plate are partly immersed in sea water and are connected through an external resistance. After a few seconds any porous spots are indicated by small bubbles of gas, and a current of some micro-amperes, which increases as corrosion increases, is recorded on an ammeter. If no bubbles are visible in 5 min. the coating is considered to be homogeneous and adhering well.—J. C. R.

Anti-Corrosion Surface Treatments and Protective Coatings. M. Ballay. (Métaux, Corrosion, Usure, 1941, vol. 16, Jan.–Feb., pp. 3–12; Mar.–Apr., pp. 19–23). Metals can be protected against corrosion by applying coatings or by modifying their surface structure. Coatings may be non-metallic, such as paints, varnishes, lacquers, and enamels, or metallic, applied by tinning, galvanizing, metallizing, cladding or electrodeposition. Each process is briefly described. The second part of the article deals with cementation processes including nitriding, Sherardizing, and chromizing, and other surface-treatment processes such as Parkerizing and Bonderizing.—J. C. R.

The Control and Maintenance of Electroplating Solutions. P. Berger. (Sheet Metal Industries, 1947, vol. 24, Sept., pp. 1820–1825). This is the concluding section of a series. (*See* Journ. I. and S.I., 1947, vol. 157, Nov., p. 473.)

Treating Cyanide Waste from Metalworking Plants. J. G. Dobson. (Third Industrial Waste Conference: Steel, 1947, vol. 121, Sept. 15, pp. 93, 112–117). Details are given of the use of chlorine for the oxidation of the cyanide in the waste from such metallurgical processes as electroplating, case-hardening, pickling, and scrubbing of blast-furnace gas. The method has proved successful in laboratory tests and large-scale plant installations.—C. O.

Contributions of Sir Humphrey Davy to Cathodic Protection. I. A. Denison. (Corrosion, 1947, vol. 3, June, pp. 295–298). A summary is given of the papers on the protection of the copper sheathing on wooden ships which Sir Humphrey Davy presented before the Royal Society.—C. O.

A Comparison of British and Spanish Practice in the Chromium Facing of Press Dies. J. B. V. Raga and J. A. Marly. (Técnica Metalúrgica : Sheet Metal Industries, 1947, vol. 24, Sept., pp. 1829–1835, 1839). The authors comment on a paper by E. A. Ollard and E. B. Smith (*see* Journ. I. and S.I., 1946, No. II, p. 76A), and quote instances of tools with a hardness greater than 500 Vickers which, after chromium plating, had a very long life. Spanish methods of plating are also contrasted with the technique described by Ollard and Smith.—R. A. R.

Automatic Electroplating Line. A. H. Allen. (Steel, 1947, vol. 121, Aug. 25, pp. 80–83, 126–128). The plant of the Pontiac Motor Division, Pontiac, Michigan, for the copper-, nickel-, and chromium-plating of motor-car parts is described and illustrated. The completely automatic processes enable 11,000 stamped steel parts, with a surface area of 45,000 sq. ft., to be plated every 16 hrs.—C. O.

Silver Plating Facilitates Bonding Glass to Steel. (Steel, 1947, vol. 121, Aug. 25, pp. 92–95). A process for bonding glass to steel by means of a fused electro-deposited silver plating is described.—C. O.

Metal Spraying—Some Notes on the Reclamation of Worn or Faulty Parts. J. Porter. (Automobile Engineer, 1947, vol. 37, Sept., pp. 343–346). Examples are given of the application of metal spraying for rebuilding metal surfaces.—C. O.

How to Maintain the Impression of Quality Despite Restrictions on the Use of Non-Ferrous Metals. P. Orlowski. (Métaux, Corrosion, Usure, 1941, vol. 16, May–June, pp. 41–43). The use of non-ferrous metallic coatings was restricted in France during the war. Brief details are given of the Parkerizing and Bonderizing processes for treating iron and steel. The products can then be coated with paint or varnish.—J. C. R.

Protecting Surface Finish with Strippable Plastic Coatings. H. R. Clauser. (Materials and Methods, 1947, vol. 26, July, pp. 70–74). The protection of metal surfaces by the application of strippable plastic coatings is discussed and the nature, characteristics, properties, and uses of (1) solvent types, (2) water-dispersion types, and (3) hot-melt types of coatings are described.—J. R.

POWDER METALLURGY

Powdered Metals and the Engineer. H. W. Greenwood. (Transactions of the Institution of Engineers and Shipbuilders in Scotland, 1946–47, vol. 90, pp. 101–115). A general account is given of the preparation, pressing, and sintering of metal powders, and of their engineering applications.—C. O.

Theory and Possibilities of Powder Metallurgy. M. Zeicher. (Revue Universelle des Mines, 1946, Series 9, vol. 2, No. 6, pp. 234–243). The theory and practice of powder metallurgy are reviewed.—R. F. F.

The Manufacture of Sintered Magnets in the "Magnetfabrik Dortmund" of the Deutsche Edelstahlwerke A.G., Krefeld. (British Intelligence Objectives Sub-Committee, 1947, F.I.A.T. Final Report No. 1130 : H.M. Stationery Office). The methods used at the Dortmund plant of the Deutsche Edelstahlwerke A.G., Krefeld, for the production of sintered magnets are described in detail and compared with those formerly used at the plant at Reutte, Tyrol.—C. O.

Small Parts Made by Powder Metallurgy. H. W. Greenwood. (Iron and Coal Trades Review, 1947, vol. 155, Sept. 26, pp. 613–614). Some of the possibilities in the mass production of small parts by means of powder-metallurgy methods are discussed, special reference being made to the use of Swedish sponge-iron powder.—J. R.

Introduction to Seminar—Review of Literature on Pressing of Metal Powders. R. P. Seelig. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2236 : Metals Technology, 1947, vol. 14, Aug.). The literature on the theory and practice of the pressing of metal powders at room temperatures is reviewed.—C. O.

Improved Engineering Properties of Parts Made from Iron Powders. C. G. Goetzel. (Product Engineering, 1947, vol. 18, Aug., pp. 115–119). The limitations in the mechanical strengths of iron powder compacts which can be attained by ordinary pressing and sintering methods are discussed. Data are presented to show the great improvements brought about by the infiltration of the pores with copper alloys and by partial alloying of the iron with copper.—C. O.

PROPERTIES AND TESTS

Purpose, Results and Future Prospects of Testing Materials in Switzerland. M. Rŏs. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 3, pp. 85–95). Swiss methods for testing materials are reviewed—R. F. F.

Recent Developments and Future Prospects in the Study of Resistance of Materials. C. Massonnet. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 4, pp. 118–126). Recent theoretical and experimental progress in the field of the strength of materials is reviewed and lines of future research are suggested.—R. F. F.

Universality of Chevenard's Differential Dilatometer with Photographic Recording. V. Mathieu. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 7, pp. 259–266). New applications in the laboratory of Soc. An. Ougrée-Marihaye of a modified form of Chevenard's differential dilatometer are described.—R. F. F.

Recent Technical Progress in the Micro-Mechanical Analysis of Metals. P. Chevenard, X. Waché, and E. Joumier. (Métaux, Corrosion, Usure, 1941, vol. 16, Jan.–Feb., pp. 13–14). Descriptions are given of micro-testing machines for measuring impact strength, creep, dynamic and static torsional strength, and internal friction.—J. C. R.

The Elastic and Breaking Limits of Hot-Rolled Round Bars. Å. Holmberg. (Teknisk Tidskrift, 1947, vol. 77, Sept. 27, pp. 724–725). [In Swedish]. Data from tests on steel bars from 4·5 to 28·5 mm. in dia. are presented which show that the ratio elastic-limit/fracture-stress decreases with increasing diameter of the bars.—R. A. R.

A Statistical Theory of Fracture. J. C. Fisher and J. H. Hollomon. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2218 : Metals Technology, 1947, vol. 14, Aug.). The size effect in solids, the scatter of fracture-stress values, and the dependence of fracture stress upon strain are analysed statistically, and an attempt is made to determine the quantitative relationship between structure and fracture stress.—C. O.

Influence of Plastic Deformation, Combined Stresses, and Low Temperatures on the Breaking Stress of Ferritic Steels. D. J. McAdam, jun., G. W. Weil, and R. W. Mebs. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2220 : Metals Technology, 1947, vol. 14, Aug.). The results are presented of tensile tests on deep-notched low-carbon steel specimens at temperatures between room temperature and —188° C., and of fracture tests at —188° C. of specimens previously plastically deformed at room temperature. The influences of plastic deformation and the stress system on the flow and fracture of ferritic steels are discussed at length.—C. O.

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Upper Yield Point: Occurrence in Bending Tests and Signification. C. Benedicks and R. Skorski. (Nature, 1947, vol. 160, Sept. 20, p. 399). Experiments which show the existence of an "upper" yield point in metallic wires under bending stresses are reported. This upper yield point is considered to be present when a hard precipitate is present on the grain boundaries.—C. O.

An Investigation of Beams with Butt-Welded Splices Under Impact. Part I—Impact and Static Tests on Beams with Butt-Welded Splices and Plain Beams Including Preliminary Studies of the Resistance to Impact at Sub-Zero Temperatures. W. J. Krefeld and E. C. Ingalls. (Welding Journal, 1947, vol. 26, July, pp. 372-s–400-s). The results are presented of experimental and mathematical analyses of the dynamic behaviour under impact loading of rolled structural beams with butt-welded splices.—C. O.

Auxiliary Tests on the Steels of I-Beams Tested in Flexural Impact at Columbia University. G. G. Luther, W. E. Ellis, and C. E. Hartbower. (Welding Journal, 1947, vol. 26, July, pp. 400-s–408-s). The results are given of notched impact and notched bend tests at various temperatures on 16-in. I-beams with butt splices, which were used in flexural impact tests during an investigation at Columbia University.—C. O.

The Problem of Determining the Toughness of Metallic Materials. A. Legat and F. Fux. (Berg- und Hüttenmännische Monatshefte der Montanistichen Hochschule in Leoben, 1947, vol. 92, June, pp. 101–104). The adequacy of the notched-bar impact test for determining the toughness of metallic materials is discussed.—R. A. R.

The Apparent Yield Strength of Plain Carbon Steel. J. A. Pope. (British Association: Engineering, 1947, vol. 164, Sept. 19, pp. 284–288). The present state of knowledge concerning the laws which govern the yield of metals under compound stress conditions is discussed. The analysis is mainly confined to the consideration of normalized plain carbon steels, and suggestions are made concerning size and stress-distribution effects which bring the various experimental results into agreement.—C. O.

Large Strains and Displacements in Stress-Strain Problems. K. H. Swainger. (Nature, 1947, vol. 160, Sept. 20, pp. 399–400). A theory is developed which gives the relationship between the stresses and any finite values of displacement and strain in a loaded structure.—C. O.

A Graphic Method of Presenting Strength Properties. A. Meldahl. (Schweizer Archiv, 1944, vol. 10, Sept., pp. 269–274). By the methods of Haigh and Westergaard it is possible to represent the three principal stresses by ordinates at right angles to each other in a three-dimensional diagram so that any conditions of stress can be represented by a point in the diagram. All states of stress resulting in equal strain for a given material will then be represented by a plane. It is proposed to make isometric projections of these planes so that their shapes will be represented by contour lines. Examples are given which demonstrate that the strength properties of both isotropic and anisotropic materials can be presented by this technique.—R. A. R.

Resistance of Materials—Theory and Practice. P. Laurent. (Génie Civil, 1947, vol. 124, Oct 1, pp. 370–373). The theoretical principles underlying the calculation of the resistance of materials to stress are outlined, and safety factors are considered.—R. F. F.

Symposium on Internal Stresses in Metals and Alloys. (Institute of Metals, October, 1947). This Symposium, which was organized by the Institute of Metals in association with the Faraday Society, the Institute of Physics, the Institution of Mechanical Engineers, the Iron and Steel Institute, the Physical Society, and the Royal Aeronautical Society, was held in London on October 15 and 16, 1947. The proceedings were divided into four sessions, dealing with: (*a*) The measurement of internal stresses; (*b*) the origin, control, and removal of internal stresses; (*c*) effects on a microscopic and submicroscopic scale associated with internal stresses; and (*d*) effects on a macroscopic scale associated with internal stresses. The number of papers presented was 36. Abstracts of those with a bearing on ferrous metallurgy follow.—C. O.

Mechanical Methods for the Measurement of Internal Stress. H. Ford. (*Ibid.*, pp. 3–11). Various mechanical methods of measuring the intensity and distribution of internal stresses in cylinders, thin-walled tubes, plates, and bars are described and the basis of the calculation explained. The paper includes a bibliography of the principal literature on the subject.—C. O.

The Investigation of Internal Stresses by Physical Methods other than X-Ray Methods. R. King. (*Ibid.*, pp. 13–23). Measurements of magnetic properties, electrical resistivity, and internal friction have been used to observe changes in internal stresses produced by treatments such as cold work, quenching, or annealing, and tests have been developed to identify material defective on account of high internal stresses. Quantitative determinations have been made by means of magnetic measurements on plastically stretched and cold-drawn nickel wires, but such determinations are limited at present to materials the magnetostriction of which may be treated as isotropic.—C. O.

Measurement of Internal Stresses by X-Rays. D. E. Thomas. (*Ibid.*, pp. 25–31). A short account is given of the theory and technique of the measurement of internal stresses in metals by X-ray methods.—C. O.

Some Fundamental Aspects of the Application of X-Rays to the Study of Locked-Up Stresses in Polycrystalline Metals. W. A. Wood. (*Ibid.*, pp. 31–34). The significance of internal-strain values as determined by X-ray technique, and of the derived values of the internal stress or stresses in relation to values determined by other methods, is discussed. When plastic deformation has occured, the X-ray technique reveals variations in residual strain on a microscopic scale. The mechanical methods of studying locked-up stresses measure only the macro effects, whereas the X-ray technique measures the net effect of macro and micro stresses.

The Principles of the Interpretation of X-Ray Photographs of Imperfect Crystals. H. Lipson. (*Ibid.*, pp. 35–44). An outline is given of the theory of the reciprocal lattice relevant to the diffraction of X-rays in imperfect crystals. This theory is applied to the discussion of recent experimental work on the nature of plastic deformation in metals and imperfect crystal structures.—C. O.

Classification and Nomenclature of Internal Stresses. E. Orowan. (*Ibid.*, pp. 47–59). The main types of internal stress are reviewed; they fall into two groups: (1) Body stresses, arising from (usually macroscopic) non-uniformities of external (mechanical, thermal, or chemical) influences acting upon the body; (2) textural stresses, due to textural (usually microscopic) inhomogeneities which may either be present in the material initially, or be produced by plastic deformation or structural changes. Attention is directed to the possibility of investigating textural stresses in transparent single crystals or polycrystalline materials by means of polarized light.

László's Papers on Tessellated Stresses: A Review. F. R. N. Nabarro. (*Ibid.*, pp. 61–72). A comprehensive summary is made of the four papers on tessellated stresses written by F. László (*see* Journ. I. and S.I., 1943, No. I, pp. 173P–201P ; 1943, No. II, pp. 137P–159P ; 1944, No. II, pp. 183P–207P ; and 1945, No. II, pp. 207P–228P).—C. O.

Internal Stresses in Steel Castings. H. Elliss. (*Ibid.*, pp. 85–93). A summary is made of the literature on the origin and effects of internal stresses in steel castings. The major factors producing, and the major effects of, internal stress are dealt with, *i.e.*, casting design, moulding technique and foundry practice, solid contraction in the mould, welding operations, fettling, pickling and heat-treatment processes, and the effects of steelmaking practice and steel composition on the susceptibility to " hot tear " formation are reviewed.

The Stresses in Large Masses of Steel Cooling from the Austenitic Region. J. E. Russell. (*Ibid.*, pp. 95–106). The distinction between tessellated and macro stresses is considered briefly, and it is pointed out that in order to estimate the latter it is essential to determine the course of the austenite transformation and the temperature at all points of the cooling body. An approximate method of deriving these data from the **S** curves which summarize the isothermal-transformation data is outlined for a eutectoid carbon steel, and the results from an 8-cm. dia. bar of such a steel are computed for two conditions of cooling.

Stress in Electrodeposited Metals. A. W. Hothersall. (*Ibid.*, pp. 107–118). Stress in electrodeposited metals is normally measured by the bending of a thin strip cathode which is plated on one side only ; various adaptations of this method are illustrated. Chromium, cobalt, copper, iron, nickel, and silver are normally deposited in a state of tensile stress (which may reach 20–40 tons/sq. in. in chromium, cobalt, or nickel), cadmium, lead, and zinc deposits are generally in compressive stress which, for zinc, continues to increase after deposition is finished. The magnitude and sometimes the direction of the stress varies with conditions of deposition.

Residual Stresses Due to Welding. R. Weck. (*Ibid.*, pp. 119–129). A short résumé is made of the causes of residual stresses in steel after welding. Measurements of plastic deformations and residual stresses in the vicinity of butt welds in mild-steel plates have been made with the Tomlinson strain gauge. It is concluded from the results that stresses of the magnitude of the yield point will always be present in welded steel. The results of experiments in which welded specimens were restrained while cooling have led the authors to express the opinion that, under static loading conditions, the strength and carrying capacity of steel will not be influenced by residual stresses if yielding takes place.—C. O.

Internal Stresses Produced by the Sliding of Metals. F. P. Bowden and A. J. W. Moore. (*Ibid.*, pp. 131–137). When a hemispherical steel surface slides over a copper surface a clearly defined track is formed. Direct examination shows little apparent difference between the track produced under clean and under lubricated conditions. However, for clean sliding, sections reveal much more severe distortion of the copper beneath the surface of the track. Study of the clean sliding of copper on steel shows that adhesion and a plucking out of the steel occurs at the interface. When the surfaces are lubricated, a similar process takes place on a reduced scale, as is shown by chemical analysis and tests with radio-tracer elements. It is concluded that the internal stresses produced below the surface of the copper result from the formation and shearing of metallic junctions during sliding. Continued sliding would give a highly worked Beilby film with a distorted layer extending far beneath the surface.

Some Internal Stresses in Turbine Rotors. M. C. Caplan, L. B. W. Jolley, and J. Reeman. (*Ibid.*, pp. 139–152). Measurements of the expansion during rotation of a special unbladed turbine wheel have been made to determine whether any permanent set occurred after first running to full speed. The sum of the centrifugal stresses and any internal stresses remaining in the wheel after it has been normalized and tempered was insufficient to produce any measurable permanent set.

Tests on turbine shafts at steam-turbine temperatures have shown that a shaft forging may remain true, may take a permanent set, or may be deflected to a degree dependent on the temperature. The stress relief necessary to give the observed permanent sets, and the difference in coefficient of thermal expansion across a shaft necessary to produce a given deflection, have been calculated.—C. O.

Residual Stresses in Beams after Bending. G. Forrest. (*Ibid.*, pp. 153–162). The distribution of residual stresses in beams bent into the plastic range is described, first for beams of rectangular section, and secondly for beams which may be considered as of **T**-section. It is shown that these stresses in certain forms of beams may approach the ultimate tensile stress of the material. Some comments on the effects of stretch bending and reverse bending are also included.

Stress-Relief Treatment of Iron Castings. (*Ibid.*, pp. 179–188). This is a report by Sub-Committee T.S. 17 of the Technical Council of the Institute of British Foundrymen. The Institute of British Foundrymen was approached by the British Internal Combustion Engine Research Association in connection with a study of stress-relief practice for iron castings. Sub-Committee T.S. 17 was formed, and reviewed the literature, information available in the private records of some members, and data supplied by a number of firms as a result of a questionnaire. Information was collected on technical aspects of the problem and current practices by firms engaged in the production of a number of types of iron castings covering turbine parts, machine tools, chemical castings, automobile castings, engine castings, etc. After a study of the data supplied, the Sub-Committee has drawn up recommendations for practice for full stress relief, and for modified stress relief for stabilizing iron castings. Due consideration is given to the possible effects of temperature on the properties of the metal. The report concludes with a bibliography of available information.

Stresses Induced by the Shot-Peening of Leaf Springs. J. C. W. Humfrey. (*Ibid.*, pp. 189–193). In connection with a series of fatigue tests on vehicle leaf springs after shot-peening under various conditions, measurements were made of the surface compressive stresses induced by the peening. The measurements were obtained by observing the change of camber of the leaves after removal of successive surface layers in hot dilute sulphuric acid. It is shown how the stress values can be obtained from the changes of camber. Some typical results showing the variation in compressive stress with depth are given and discussed. Maximum stress values of 35–60 tons/sq. in.. and a

compressively stressed layer of from 0·006 to 0·016 in. deep were found.

The Production of Favourable Internal Stresses in Helical Compression Springs by Pre-Stressing. D. G. Sopwith. (*Ibid.*, pp. 195–207). The load-carrying capacity of a helical compression spring can be increased considerably by pre-stressing or "scragging," *i.e.*, compressing the spring (wound longer than required) to closure several times. This results in the introduction of favourable internal or residual shear stresses. The principles underlying the process are investigated and the distribution of the residual stresses and of the stresses under load are derived. Examples of the application of these principles to various problems arising in the design of scragged springs are given.

Autofrettage. A. G. Warren. (*Ibid.*, pp. 209–218). An account is given of the present position of the theory of autofrettage.—C. O.

Effects Associated with Stresses on a Microscopic Scale. Sir Lawrence Bragg. (*Ibid.*, pp. 221–226). The paper attempts to establish a formula for the limit at which a metal when sheared ceases to behave elastically and begins to undergo plastic deformation. The formula is based on the consideration that a slip process confined to a limited volume of the metal should on the whole represent a local release of energy. The formula yields estimates of the elastic limit which are of the same magnitude as those actually observed in cold-worked pure metals.

Internal Stresses Arising from Transformations in Metals and Alloys. F. C. Thompson. (*Ibid.*, pp. 227–232). The literature on the stresses set up by transformations in metals and alloys is reviewed.—C. O.

A Note on the Effect of Internal Stresses on the Rates of Transformation in Iron–Nickel Alloys. C. C. Earley. (*Ibid.*, pp. 233–236). During the transformation $\gamma \rightarrow \alpha$ on continuous rapid cooling in binary iron–nickel alloys containing up to 27% of nickel, the α-phase which forms has a distorted body-centred cubic structure. The variation in lattice strain with nickel content and its removal by suitable annealing treatments has been investigated, and the progress of the isothermal transformations $\alpha \rightarrow \gamma$ in strained and strain-free material compared. The rate of transformation does not increase with increase of lattice strain.

Diffusion and Precipitation in Alloys. F. R. N. Nabarro. (*Ibid.*, pp. 237–250). A general review is given of recent work on diffusion and precipitation in alloys, with special emphasis on the theoretical mechanism of diffusion, and on the factors governing the shape and size of the precipitate. The effects of internal stresses on diffusion and precipitation, and of precipitation on the mechanical properties, are discussed.

Age-Hardening. Marie L. V. Gayler. (*Ibid.*, pp. 255–264). A brief résumé is given of the theories of age-hardening which have been put forward during the past 26 years. Recent metallographic analyses have led to the deduction that highly localized stresses are formed within the lattice during the process of age-hardening, on a sub-microscopic as well as on a microscopic scale. The existence of such microscopic stresses is also indicated by X-ray diffraction photograms. The development during age-hardening of such stresses at highly localized points within the lattice must have a profound effect on the behaviour of the alloys when subjected to different tests, and particular emphasis is laid on the importance of this fact in relation to stress-corrosion and creep tests.

Internal Stresses and the Formation of Hair-Line Cracks in Steel. J. H. Andrew and Hsun Lee. (*Ibid.*, pp. 265–273). The formation of hair-line cracks in steel is discussed and explained on the basis of internal pressures caused by hydrogen. It is suggested that molecular hydrogen could be accommodated at the mosaic disjunctions or boundaries and that the variation of the size of the mosaic blocks may account for the different response of steels of different compositions and treatments to hair-line-crack formation.

The Experiments of Boas and Honeycombe on Internal Stresses Due to Anisotropic Thermal Expansion of Pure Metals and Alloys. F. P. Bowden. (*Ibid.*, pp. 275–280). The surface cracks and slip lines which are formed when non-cubic metals and alloys are subjected to repeated cycles of heating and cooling have been attributed to stresses caused by anisotropic thermal expansion. Several experiments which support this hypothesis are described. The effect of the anisotropic expansion on stresses during casting and annealing are discussed.—C.O.

A Photoelastic Approach to Stress Modifications Caused by Inhomogeneities. B. Sugarman. (*Ibid.*, pp. 281–288). Details are given in this paper of the technique adopted to investigate, by means of photo-elastic beam samples, the stress concentrations around holes and patterns of holes in isotropic metallic bodies. The investigations have primarily been carried out with two-dimensional sensitive plastics, but have been supplemented by three-dimensional investigations.

Influence of Residual Stress on Chemical Behaviour. U. R. Evans. (*Ibid.*, pp. 291–310). Descriptions are given of experiments in which oxide films from nickel were transferred to vaseline-coated glass. The thicker films wrinkled, suggesting uniform stresses due to misfit between the volumes occupied by oxide and metal; the thinner films curled into rolls, suggesting stress gradients, probably inherited from surface stresses in the metal.

Work on the subjects of the effect of residual stresses on wet corrosion, stress-corrosion cracking, and the prevention of corrosion is reviewed.—C. O.

Chemical Manifestations of Internal Stress. F. H. Keating. (*Ibid.*, pp. 311–332). The combined effect of internal stress and a corrosive environment is discussed in a general way, and the significance of this combination—generally referred to as stress-corrosion cracking—is indicated. The literature on stress-corrosion cracking of the commoner industrial alloys is briefly reviewed, and some examples of such cracking in industrial plant are recorded. The various factors involved in stress-corrosion cracking are considered, and a tentative explanation of the mechanism of such cracking is proposed.

Note of Stress-Corrosion Cracking of Steels in the Presence of Sulphur Compounds. W. P. Rees. (*Ibid.*, pp. 333–335). Reference is made to cases of failures of ferrous materials by stress-corrosion cracking, where the corroding agent was probably hydrogen sulphide or some sulphur-containing compound. Laboratory experiments are referred to which have shown that hardened and tempered alloy steels used for gas-cylinder manufacture are susceptible to stress-corrosion cracking in the presence of moist hydrogen sulphide or carbon disulphide.

Internal Stresses in Railway Materials. H. O'Neill. (*Ibid.*, pp. 337–349). Eighteen examples of the occurrence of internal stress are outlined, in half of

which it is deliberately employed to play a useful part. Details are given of manufacturing stresses in various types of rail, and the effect of high speed pounding by locomotives is mentioned. Values have been determined of the stress in peened piston rings, shot-blasted spring plates, boiler rivets, and expanded boiler tubes. The effect of service straining on stress removal is discussed. Fatigue cracking in axle press-fits is outlined, and the shrink-fits of wheels and tyres are considered in relation to thermal effects from braking and welding.

Delayed Cracking in Hardened Alloy Steel Plates. E. H. Bucknall, W. Nicholls, and L. H. Toft. (*Ibid.*, pp. 351–365). After a brief résumé of a war-time investigation into cracking in hardened alloy-steel plates and welded structures, for which the responsibility lay with internal stresses, the authors describe the "conical disc test" which was developed as a means of simulating internal stresses in steels of this class, so that the effects of various factors on the cracking phenomenon could conveniently be studied. The application of this test on a statistical basis led to the recognition of the dependence of cracking susceptibility on composition and on tempering treatment.

The Stress System Causing Hard-Zone Cracking in Welded Alloy Steels. J. A. Wheeler. (*Ibid.*, pp. 367–373). The broad features of the problem of hard-zone cracking in welded alloy steels are reviewed, and the mechanical properties of the metal which cracks are discussed in relation to the stresses which act on it. It is shown that cracking can be accounted for by the combined action of macro- and micro-stresses on a material embrittled by heating to high temperature during welding.

The Strength Properties and Microstructure of Arc Welds in Very Thick Unalloyed Structural Steels. R. Montandon. (Schweizer Archiv, 1945, vol. 11, Apr., pp. 97–103; May, pp. 147–158). Comprehensive tests were made on welded joints in three types of steel plate 80 to 100 mm. thick. The three types contained 0·09%, 0·25%, and 0·44% of carbon respectively. In most cases specimens with a gauge length of 7 mm., 1·5 mm. in dia., were cut from the weld metal, the boundary zone, and the parent metal. The microstructures and mechanical properties at normal and at elevated temperatures are compared. The tests proved that, by employing a good technique with correct electrodes, perfect welded joints can be made in all three types of steel even up to 100 mm. in thickness.—R. A. R.

Testing Gun Steel and Other Alloys and Metals for Resistance to Surface Cracking. E. Ingerson. (American Institute of Mining and Metallurgical Engineers: Metals Technology, 1947, vol. 14, Aug.). An account is given of a method devised for determining the resistance of metals and alloys to surface cracking under physical and chemical conditions similar to those present in the bores of guns.—C. O.

A Twelve-Channel Recorder for Use with Resistance Strain Gauges. A. Watson. (Journal of Scientific Instruments, 1947, vol. 24, Sept., pp. 239–242). A detailed description is given of the design of an instrument for recording the change in electrical resistance of strain gauges attached to a ship's hull. The gauges are included in two arms of an A.C. bridge circuit; the off-balance voltage is amplified and rectified, and the deflection produced by it on twelve cathode-ray tubes is continuously recorded photographically on one film. Transient strains of 0·1 sec. duration can be recorded.—C. O.

New Hardness Tester. T. Dobry. (Hutnické Listy, 1946, vol. 1, pp. 177–180: Industrial Diamond Review, 1947, vol. 7. Aug., pp. 244–245). Illustrated particulars are given of the new Poldi hardness tester, similar to the Firth-Brown Hardometer. In the new instrument the impression is made mechanically, and measurements are made optically, so that the image of the impressions can either be magnified or projected on to a screen. The pressure mechanism is spring-loaded and a pressure of 30 kg. is recommended in using the machine for all metals including hardened steel.—R. E.

T-Band Method Tailors Steel to Function of Part. P. R. Wray and R. W. Roush. (S.A.E. Journal, 1947, vol. 55, Aug., pp. 17–20). A simple explanation is given of the uses and advantages of S.A.E.-A.I.S.I., hardenability specifications for steels.—C. O.

Surface Finish. F. C. Johansen. (Indian Engineering, 1947, vol. 121. May, pp. 219–221; vol. 122, July, pp. 25–28). Methods for studying the physical properties and measuring the roughness of highly finished surfaces are enumerated.—C. O.

Magnetic Development in Japan During World War II. (British Intelligence Objectives Sub-Committee, 1947, Report No. B.I.O.S./J.A.P./P.R./1283: H.M. Stationery Office). Japanese wartime achievements in research on magnetism and the development and application of magnetic alloys are summarized.—C. O.

The Reluctometer. S. Bemert. (Teknisk Tidskrift, 1947, vol. 77, June 14, pp. 523–524). [In Swedish]. A brief description is given of an instrument for checking the uniformity in the quality of mass-produced tools and other steel parts; the instrument measures the magnetic reluctance of the steel.—R. A. R.

Graphitization of Steels at Elevated Temperatures. A. E. White. (Metal Progress, 1947, vol. 52, Sept., pp. 371–375). A short account is given of the phenomenon of graphitization in steels at high temperatures; it is based on the results of numerous recent investigations.—C. O.

Metallic Creep. A. H. Sully. (Research, 1947, vol. 1, Oct., pp. 19–24). A general review is made of the subject of metallic creep—its mechanism and metallurgical significance.—C. O.

Anomalous Changes in Tensile Properties of Quenched Iron–Cobalt (35 Pct Co) Alloys. J. K. Stanley. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2221: Metals Technology, 1947, vol. 14, Aug.). The mechanical properties of 14 iron-cobalt alloys containing 35% of cobalt and various minor additions of vanadium, molybdenum, chromium, etc., have been determined after quenching in water at 0° C., from temperatures up to 1000° C. Specimens quenched from above 600° C. were found to be less hard than those quenched below 600° C. Tensile strength and yield strength decreased on quenching from increasing temperatures up to the transformation range, above which they increased. Elongation reached a maximum on quenching from 850–900° C. Metallographic examination showed that the grain-size was refined by quenching from above the critical temperature, but no precipitation was noted after quenching and ageing.—C. O.

The Rôle Played by Superficial Layers of Metal in the Phenomenon of Pickling Embrittlement of Extra-Soft Annealed Steel. P. Bastien. (Comptes Rendus, 1941, vol. 212, Apr. 28, pp. 706–708: Métaux, Corrosion, Usure, 1941, vol. 16, Mar.–Apr., pp. 28–29). Brief details are given of experiments carried out on extra-soft annealed steel wires up to 2 mm. in dia. to examine the mechanism of pickling embrittlement.—J. C. R.

On the State and the Diffusion of Hydrogen in Pure Iron at Ordinary Temperatures (Pickling Embrittlement). G. Chaudron and L. Moreau. (Académie des Sciences, Nov. 24, 1941 : Métaux, Corrosion, Usure, 1941, vol. 16, Dec., pp. 124–125). Pure iron, degassed by electronic bombardment, was allowed to absorb hydrogen electrolytically in a sulphuric acid solution and also by pickling. The depths of diffusion are compared.—J. C. R.

Application of the Method of Degassing by Electronic Bombardment to the Determination of Gases in Nickel–Chromium Steels. L. Moreau. (Académie des Sciences, Nov. 24, 1941 : Métaux, Corrosion, Usure, 1941, vol. 16, Dec., pp. 125–126). Results are presented of tests carried out by electronic bombardment to determine the gas content of various alloys, including nickel–chromium steel. The results obtained for nitrogen agreed closely with those obtained by chemical analysis. —J. C. R.

Influence of Cold Work on the Diffusion of Atomic Hydrogen in Extra-Soft Steel. P. Bastien. (Comptes Rendus, 1941, vol. 212, May 12 : Métaux, Corrosion, Usure, 1941, vol. 16, Mar.–Apr., pp. 29–30). Tests were carried out on extra-soft steel wire 2 mm. in dia. to compare the volume of hydrogen absorbed, under similar conditions, in the annealed and cold-drawn states, and to study the influence of preliminary annealing on the volume of hydrogen released, and the rôle of cold-drawing as a time factor on the decrease in pickling embrittlement.—J. C. R.

Effect of Sulphur in Cast Steel. G. A. Lillieqvist. (American Institute of Mining and Metallurgical Engineers : Electric Furnace Steel Conference, 1946, vol. 4, pp. 219–230). A review is made of the results of several investigations on the influence of sulphur on the properties of cast steel.—C. O.

Possibilities in the Use of Sodium. C. N. Arnold. American Institute of Mining and Metallurgical Engineers : Proceedings of the Electric Furnace Steel Conference 1946, vol. 4, pp. 248–250). Very brief mention is made of the improvement in impact strength obtained in steels desulphurized by direct additions of sodium. No details are given of the process.—C. O.

Effect of Selenium Additions to Cast Steel. T. N. Armstrong. (American Institute of Mining and Metallurgical Engineers : Electric Furnace Steel Conference, 1946, vol. 4, pp. 250–254). Details are given of laboratory and commercial tests to determine the influence of selenium on the ductility of cast steels containing sulphur.—C. O.

Selenium Additions to Cast Steel. A. P. Gagnebin. (American Foundryman, 1947, vol. 12, Aug., pp. 43–52). The influence of selenium on the distribution and form of sulphide inclusions in cast high-sulphur nickel–manganese steel is shown in a series of photomicrographs. Extensive tests in laboratory and commercial heats indicate that calcium–selenium additions enable steels of higher ductility to be obtained than when other methods of deoxidation are used.—C. O.

The Evolution of the Standard Rail of the Belgian Railways. C. Lemaire. (Revue Universelle des Mines, 1945, Series 9, vol. 1, Dec. 15, pp. 213–219). The author suggests improvements that can be made to the 50 kg. rail at present in use in Belgium. The proposed rail would not differ greatly from the present one and would allow speeds of 150 km./hr. and a maximum weight of 24 to 25 tons/axle.—R. F. F.

Effect of Stabilizing and Stress Relief Heat-Treatment for 18–8 Stainless. W. G. Hubbell. (Steel, 1947, vol. 121, Aug. 25, pp. 86–88, 114–119). Photomicrographs taken during an investigation of the effect of stabilizing and stress-relieving heat-treatments on titanium- and columbium-treated 18/8 stainless steels are reproduced. The treatments are considered to result in slight improvements in resistance to corrosion in electrolytes.—C. O.

Stainless Steels and the Chemical Industries. P. Bastien. (Métaux, Corrosion, Usure, 1941, vol. 16, July–Aug., pp. 53–58 ; Sept., pp. 67–74). The first part of this paper deals with the main features of the structure of stainless steels, the influence of increasing the nickel content and the effect of carbides on the corrosion resistance of stainless steels. In the second part the principal types of stainless steels and their properties are considered, with special reference to chromium steels, temper-hardened, having a pearlitic structure in the annealed state and martensitic structure in the tempered state, and not temper-hardened, having a ferritic structure, and chromium–nickel austenitic steels. The third part deals with the special requirements of stainless steels for use in the chemical industries. —J. C. R.

Study of the Rupture of Joint Sleeves in Overhead Electric Lines. E. H. Hubert. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 9, pp. 342–348). A study of the causes of the rupture of steel–aluminium overhead lines in the immediate neighbourhood of the sleeve, showed that, owing to a defective contact in the aluminium joint, an important fraction of the current passed through the steel, developing a considerable amount of heat which altered the mechanical properties of the steel and led to oxidation. Remedies are discussed.—R. F. F.

Selecting Spring Materials. F. P. Zimmerli. (Steel, 1947, vol. 121, Aug. 11, pp. 78–79, 108–128). The hot- and cold-forming properties of a variety of ferrous and non-ferrous flat and round wire materials are considered in relation to their applications.—C. O.

New High-Speed Steels. R. Mitsche, E. M. Onitsch, and W. Blüthgen. (Berg- und Hüttenmännische Monatshefte der Montanistischen Hochschule in Leoben, 1947, vol. 92, Jan.–Mar., pp. 13–21). The microstructure, hardness, and cutting properties of high-speed steels containing up to 2·6% of tungsten with small amounts of chromium, molybdenum, vanadium, and aluminium are presented and discussed, with special reference to reducing the tungsten and molybdenum, and the effect of aluminium. Additions of 0·4–0·6% of aluminium with a carbon content of 1·1% max. gave the best results.—R. A. R.

Some Engineering and Design Considerations in the Utilization of Ordinary and High-Strength Steels. W. B. Brooks. (Welding Journal, 1947, vol. 26, Aug., pp. 665–669). The author discusses the functions of mechanical testing, particularly the notched impact test, in the selection of steels for engineering structures. It is maintained that many brittle-type failures are due to faulty design and construction and that efficient design can enable full advantage to be taken of the high yield points of high-strength steels in the utilization of sections of reduced area and capable of withstanding severe loading without brittle failure. The influence of plate thickness and fatigue upon design are also considered.—C. O.

High Strength Cast Irons for Gears. E. M. Currie. (Machinery, 1947, vol. 71, Sept. 11, pp. 291–295). The physical and chemical properties of ordinary cast iron and Meehanite metal are discussed and compared with reference to their use in the manufacture of gears.—J. R.

A Theoretical Discussion of Pitting Failures in Gears. R. Beeching and W. Nicholls. (Institution of Mechanical Engineers, 1947, Preprint). The stresses which cause the failure of the surface of gear teeth are considered theoretically. Methods of calculating the permissible line load between contacting cylindrical surfaces for static or cyclic conditions are indicated, and the effects of friction are discussed. Suggestions are made for the maximum permissible case thicknesses for case-hardened steel components subjected to contact stresses of the type encountered in gears, rollers, and cams. The derivation from Hertz's general solution of expressions for the principal stresses at any point in the neighbourhood of the zone of contact between parallel cylinders is given in an appendix.—C. O.

Ferro-Nickels with Reversible Magnetic Transformation. P. Mulquin. (Echo des Mines, 1946, Apr., pp. 54–55 ; May, pp. 70–71 ; June, pp. 86–87). The elasticity, expansion, magnetic and electrical properties of ferro-nickels are discussed and examples given. The special properties of these ferronickels makes them suitable for various purposes in the electrical and radio industry. —R. F. F.

A New Method of Non-Destructive Spot-Weld Testing for Stainless Steels. A. M. Armour. (Metallurgia, 1947, vol. 36, Sept., pp. 273–275). An account is given of the use of the Metroflux captive-fluid magnetic cell for the non-destructive testing of spot welds in stainless steels.—C. O.

Creep and Fatigue as Affected by Grain Boundaries. C. Crussard. (Société Française de Métallurgie : Metal Treatment, 1947, vol. 14, Autumn Issue, pp. 149–160). An account is given of a study of the rôle of grain boundaries in creep and fatigue (particularly in the hexagonal metals, zinc and magnesium), by experimental observations and theoretical calculations. Among the conclusions reached are : (1) In pure metals in the annealed state there is no amorphous layer at the grain boundaries. (2) For very small loads there exists a special type of deformation, known as micro-creep. (3) Beyond the limit of micro-creep there appears creep of the normal type, with strain-hardening. This hardening has two origins, one intergranular, and the other intracrystalline. (4) Creep forms a distorted area, at least 0·01μ thick, around the boundaries, which may be the centre of viscous deformation through individual movements of atoms. (5) Another type of creep, of intercrystalline origin, exists at constant creep rates, because of the annealing of blocked sliding motions.—C. O.

Photographic Aspects of the Radiography of Welded High Pressure Vessels. L. Mullins. (Transactions of the Institute of Welding, 1947, vol. 10, Oct., Supplement, pp. 11–20). The desirable characteristics of the photographic materials used in the radiography of welded pressure vessels are discussed, and recommendations are made on the handling and processing of such materials. The causes of confusing shadows on radiographs are tabulated and illustrated.—C. O.

Japanese Use of High Temperature Alloys in Rocket, Jet and Gas Turbine Applications. (British Intelligence Objectives Sub-Committee, 1947, Report No. B.I.O.S./J.A.P./P.R./887 : H.M. Stationery Office). The chemical compositions and mechanical properties of the steels used by the Japanese in jet-engines and rockets are tabulated.—C. O.

Stainless Steels. (Echo des Mines, 1946, July, pp. 103–104). Austenitic steels containing 18–20% of chromium and 8–10% of nickel are studied with regard to their properties, resistance to corrosion, forgeability, welding, and uses. A list of French steels and manufacturers is appended.—R. F. F.

Cast Hard-Metal as a Wear-Resistant Material. E. L. Baay. (Metalen, 1947, vol. 2, Sept., pp. 1–6). [In Dutch]. The properties of cast tungsten carbide are discussed and a centrifugal method of casting it is described. —R. A. R.

Compound Steel–Aluminium Overhead Lines for Tramways and Trolleybuses. L. Albert. (Revue de l'Aluminium, 1946, vol. 23, Jan., pp. 14–23). Owing to the shortage of copper in France, tests have been made on the use of special aluminium-coated cables as overhead lines on trolleybus services. The method of determining the composition of a compound steel–aluminium wire which would be suitable for replacing existing copper wires is outlined. It is not stated by what means the steel and aluminium are joined.—J. C. R.

Metallurgical Examination of a Japanese Aircraft 12·7 mm. "Browning" Machine Gun. (British Intelligence Objectives Sub-Committee, 1947, Report No. B.I.O.S./J.A.P./P.R./1439 : H.M. Stationery Office).

Progress Report on Examination of Enemy Material : Metallurgical Examination of a Japanese Sakae-21 Aircraft Engine. (British Intelligence Objectives Sub-Committee, 1947, Report No. B.I.O.S./J.A.P./P.R./1467 : H.M. Stationery Office).

Metallurgical Examination of Small Arms Barrels. (British Intelligence Objectives Sub-Committee, 1947, Report No. B.I.O.S./J.A.P./P.R./1513 : H. M. Stationery Office). Data from the examination of 15 German and 5 Japanese small-arms barrels are tabulated.—C. O.

METALLOGRAPHY

Electron Microscopy. R. A. Scott. (Science Progress, 1947, vol. 35, Oct., pp. 638–651). Modern electron-microscope technique is described generally with illustrations.—M. A. V.

Electron Micrographs of Steel Fractures. H. Mahl. (Metallforschung, 1947, vol. 2, June, pp. 186–188). Brief details are given of a process for obtaining an aluminium oxide replica of the surface of a fractured steel specimen ; the replica can then be used in the electron microscope.—R. A. R.

Improvements in Electrolytic Methods of Determining Silica Inclusions in Steel. A. Bielanski. (Hutnik, 1946, vol. 13, Feb., pp. 102–105). [In Polish]. The results of electrolytic determinations of silica inclusions, using weak citric acid solutions and copper electrodes are presented. This method shows some advantages over the more usual sodium chloride solution and mercury electrodes.—W. J. W.

Optical Phenomena with Strongly Etched Metal Foils. W. G. Burgers and J. Weijdema. (Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 4, pp. 126–128). This is a French translation of the article in Dutch published in : Metalen, 1946, vol. 1, Sept. 15, pp. 6–7. (*See* Journ. I. and S.I., 1947, vol. 156, May, p. 141).—R. F. F.

The Rôle of Crystallography in the Origins of Metallography. A. Portevin. (Métaux, Corrosion, Usure, 1941, vol. 16, Oct., pp. 88–89). A brief historical account is presented of the work of Tschernoff and Osmond on crystallography.—J. C. R.

Determining Crystal Orientation by X-Rays. P. Coheur and J. M. Lejeune. (Revue Universelle des Mines, 1945, Series 9, vol. 1, July 15, pp. 28–32). A method of determining crystal orientation, particularly adaptable to large grain-size is described and results obtained given. A special type of recrystallization found in metals with face-centred cubic lattices, is examined. —R. F. F.

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The Use of X-Rays in the Study of Metals. M. Moray. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 155–163). The applications of X-rays to metallurgy are described under the general headings of radiography and spectroradiography. —R. F. F.

A Micro Examination of Eight Steels for the Inclusions Committee. J. H. Whiteley. (Journal of the Iron and Steel Institute, 1947, vol. 157, Sept., pp. 89–97). An investigation for the Inclusions Sub-Committee of the Ingots Committee of the British Iron and Steel Research Association, of the number, distribution, and nature of the inclusions in a series of eight low-alloy steels is described. In determining the number of inclusions, counts were made on both lengthwise and crosswise surfaces, it having been previously ascertained from the former that inclusions not well embedded are liable to be removed in polishing. The way this conclusion was arrived at is shown. Full details of the counting procedure are presented. The results indicate a direct relationship between the number of inclusions in the steels, and their combined sulphur and oxygen contents. A description of the types of inclusions observed in each of the steels is then given, their grain-sizes and macrostructures being also illustrated. In an appendix by G. Murfitt, an X-ray examination of acid-extracted inclusions in these steels is recorded.

On Interference Microscopy. Sir T. Merton. (Proceedings of the Royal Society, Series *A*, 1947, vol. 191, Sept. 26, pp. 1–6). A short account is given of some problems in the technique of interference microscopy, which provides a method for increasing the visibility of colourless specimens differing slightly in refractive index from the surrounding medium by enclosing them between semi-reflecting surfaces.—C. O.

Microradiography. R. Pospíšil. (Hutnické Listy, 1947, vol. 1, No. 9, pp. 193–197). [In Czech]. The authors describe the principles and methods of applying microradiography to the study of very thin polished specimens.—R. A. R.

The Diffusion Rates for Carbon in Austenite. F. E. Harris. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2216: Metals Technology, 1947, vol. 14, Aug.). Equations are developed to express the concentration gradients in the diffusion of carbon into austenite under both steady and unsteady conditions, in order to support the author's contention that the expression obtained by the Matano method, which gives a relationship between *D*, the diffusion coefficient, and the numerical values of the concentration gradients, for conditions of unsteady flow, is incorrect. —C. O.

Quantitative Metallography Point-Counting and Lineal Analysis. R. T. Howard and M. Cohen. (American Institute of Mining and Metallurgical Engineers, Technical Pubication No. 2115: Metals Technology, 1947, vol. 14, Aug.). The literature on the methods used by petrographers for estimating the relative proportions of microconstituents is reviewed. Point-counting and the Hurlbut counter are compared in their application to martensite-austenite structures. Both methods have been shown experimentally to be reliable, the linear method using the Hurlbut counter being somewhat superior to a real analysis by point counting.—C. O.

The Effect of Cobalt on the Rate of Nucleation and the Rate of Growth of Pearlite. M. F. Hawkes and R. F. Mehl. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2211: Metals Technology, 1947, vol. 14, Aug.). Details are given of experimental studies of the effect of cobalt on the transformations in carbon steels. Measurements were made of the rates of nucleation, the pearlite spacing, the rates of growth of pearlite, the rates of diffusion of cobalt and carbon in austenite, and the effect of nitrogen on transformation rates in cobalt steels. It is concluded that: (1) Although cobalt increases the rate of austenite decomposition at temperatures between $A_{equ.}$ and Ar_2, there is no change in the shape of the **S** curve or in the nature of the products; (2) the influence of cobalt is not exerted by indirect means involving such factors as austenite grain-size or the presence of second phases or inhomogeneities; (3) the rate of nucleation and the rate of growth of pearlite are increased by cobalt to the same degree; and (4) no effect of cobalt on the diffusion of carbon in austenite could be observed. —C. O.

The Anisotropy of Strip. P. Coheur and M. Choul. Revue Universelle des Mines, 1945, Series 9, vol. 1, (Nov. 15, pp. 189–192). A new type of anisotropy observed in metals with face-centred cubic lattices is discussed.—R. F. F.

Metal Crystals and Metal Strength. E. N. da C. Andrade. (Proceedings of the Royal Institution, 1947, vol. 33, Part II, pp. 237–250). The mechanism of slip in "perfect" crystals, single crystals, and polycrystalline aggregates is illustrated in order to show the nature of the strength of metals.—C. O.

Martensitic Transformation during Stamping. P. Coheur and J.-M. Lejeune. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, p. 188). The authors examined 18/8 stainless steel sheets after stamping and found that an austenite-martensite transformation had taken place.—R. F. F.

CORROSION

Ferrous Metal Corrosion and *p*H. J. Jackson. (Power and Works Engineering, 1947, vol. 42, Oct., pp. 301–304). A short outline of the theory of electrochemical corrosion is given and applied to explain the basis of the principal counter-corrosion measures.—C. O.

Metal-Protection Methods Dependent on Present Conditions. B. Thiede. (Technik, 1947, vol. 2, May, pp. 214–216). After the products of corrosion have been removed from metal parts recovered from debris it is necessary to prevent a renewal of the corrosion. Some methods of doing this, which are suitable for present conditions in Germany, are described.—R. A. R.

The Corrosion of Metals: Study of Means of Protection and Estimation of Results. N. Goldowski. (Métaux et Corrosion, 1940, vol. 15, July–Aug., pp. 57–63). The conditions required for metals to be resistant to corrosion are examined and methods of protection are briefly discussed.—J. C. R.

The Kinetics of Oxide Formation on Metals and Alloys. E. A. Gulbransen. (Electrochemical Society, 1947, Preprint No. 91-29). The influence of the following eleven variables on the kinetics of the oxidation process has been studied: Time, temperature, pressure, surface preparation (including passivation treatments), gaseous occlusions in the metal lattice, surface area, crystal orientation, gas flow, exposure to temperature cycles, reduced pressure, and stability of the oxide film. The experiments were carried out on iron, 18/8 steel, and a number of non-ferrous metals. The metals and alloys were found to follow the parabolic rate law over a certain range of time and temperature, deviations being observed at high temperatures, however. The transition rate theory of gas–metal reactions has been applied in the study of the behaviour of the various metals and alloys, the energies of activation varying

from 22,600 cal./mol. for iron to 45,650 cal./mol. for tungsten, and the entropies of activation from −28·7 for 18/8 steel to + 11·2 for tungsten. The effect of these factors on the rate of oxidation is discussed.—D. R. S.

An Electron Diffraction Study of Oxide Films Formed on High-Temperature Oxidation Resistant Alloys. J. W. Hickman and E. A. Gulbransen. (Electrochemical Society, 1947, Preprint No. 91-32). The oxides formed on a number of ferritic and austenitic alloy steels (Alchrome-6, Worthite, stainless steels 301 and 446, and alloy S-588) in the temperature range between 300° and 900° C. were studied by means of X-ray diffraction, the photographs being taken at the temperatures of formation, in order to exclude the possibility of changes occurring on cooling. The structures found were plotted as functions of time and temperature, and the attempt made to correlate them with thermodynamic data reported by other workers. Over the whole temperature range studied the oxides were found to consist either of Cr_2O_3 or a spinel-type structure.—D. R. S.

Designing to Prevent Corrosion. R. B. Mears and R. H. Brown. (Corrosion, 1947, vol. 3, Mar., pp. 97–118). A short account is given of the mechanism of electrochemical corrosion, as an introduction to a discussion of methods of preventing or reducing to the minimum the corrosion of metallic parts in service.—C. O.

Results of Some Studies of the Condensate Well Corrosion Problem. W. F. Rogers and H. E. Waldrip. (Corrosion, 1947, vol. 3, Mar., pp. 121–138). Investigations designed to determine the ability of the liquid hydrocarbons produced in condensate wells to form protective films on metal equipment are described. It was found that none of the eight condensates tested had sufficient wetting power to displace water films from steel, and that such protection as was afforded was due only to the "mass" effect, *i.e.*, the relatively large ratio of condensate to water.—C. O.

Use of Corrosion Inhibitors in Products Pipelines—A Survey of Practices. Ivy M. Parker. (American Petroleum Institute: Corrosion, 1947, vol. 3, Apr., pp. 157–168). A short summary is made of the data obtained from a questionnaire which was sent to twenty oil companies using pipelines for their products. The questionnaire, which is reproduced together with the tabulated answers in appendices, was designed to secure information concerning the use of corrosion inhibitors in pipelines.—C. O.

The Determination of Pipe Protection by the Continuous Polarity Method. W. E. Huddleston. (Corrosion, 1947, vol. 3, July, pp. 325–330). Examples are given of the location and removal of anodic areas on pipelines protected cathodically.—C. O.

Cathodic Protection Rectifiers. W. L. Roush and E. I. Wood. (Corrosion, 1947, vol. 3, Apr., pp. 169–172). Short descriptions are given of air-cooled, fan-cooled, and oil-immersed rectifiers used for the cathodic protection of pipelines.—C. O.

Failures of Domestic Hot Water Storage Tanks. C. P. Hoover. (American Water Works Association: Corrosion, 1947, vol. 3, Apr., pp. 185–191). The cause of the failure of galvanized-steel domestic hot-water tanks is discussed in the light of recent literature on the subject.—C. O.

Corrosion of Galvanized Hot Water Storage Tanks. J. M. Bialosky. (Journal of the American Water Works Association, 1946, vol. 38, Sept., pp. 1012–1020: Corrosion, 1947, vol. 3, Apr., pp. 192–197). The electrolytic corrosion of galvanized hot-water storage tanks is considered from the points of view of quality of steel and zinc, rate of flow and rate of heating of water, quality of the water and such external factors as the presence of copper or brass pipes setting up galvanic couples. Recommendations are made for minimizing corrosion.—M. A. V.

Corrosion in the Tropics. K. G. Compton. (Electrochemical Society, 1947, Apr., Preprint 91-35). A short account is given of the climatic conditions encountered in the tropical regions of the Southwest Pacific, Asia, West Africa, and Central America, and of the types of corrosion found on military equipment exposed under such conditions.—C. O.

Plastic Coatings to Control Metal Corrosion—A Review. S. P. Wilson. (Corrosion, 1947, vol. 3, Mar., pp. 141–148). A general survey is made of the applications of plastic coatings for metal protection against corrosion.—C. O.

Rubber Linings and Coatings. J. J. McNeill. (Corrosion and Material Protection, 1947, Mar.–Apr., vol. 4, pp. 13–14). A short account is given of the use of rubber for the protection of metals against corrosion and abrasion.—C. O.

Use of Glycerine in Metal Protection. Georgia Leffingwell and M. A. Lesser. (Corrosion and Material Protection, 1947, vol. 4, May–June, pp. 12, 22–23). A summary is made of the applications of glycerine in the protection of metallic surfaces. Forty-three references to the literature are given.—C. O.

Experiments on the Effects of an Electrostatic Field on the Corrosion of Steel. S. Eisner. (Revue Universelle des Mines, 1946, Series 9, vol. 2, No. 1, pp. 23–25). The study of the appearance of a passive area round a rust spot, tends to indicate the formation of an electric field between this area, acting as cathode and the rust spot acting as anode. When an artificial anode is used, there is still passivity near the anode, the degree of passivity varying with the distance from it.—R. F. F.

Results of Studies Carried Out in the Field of the Protection of Steel against Corrosion. M. Van Ryssel-berge. (Ossature Métallique, 1945, vol. 10, July-Aug., pp. 147–155). The phenomenon of corrosion is briefly reviewed and methods of protection are discussed. The requirements of paints used for this purpose are examined and classified in three groups—physical and chemical properties, and mechanical tests. Results are presented of tests on painted specimens carried out by the Commission Belge pour l'Etude de l'Acier contre la Corrosion and details are given of research in America on artificial ageing.—J. C. R.

Long-Term Corrosion Tests on Various Constructional Steels in the Open Air and in Different Sea and River Waters. (Ossature Métallique, 1947, vol. 12, Apr., pp. 167–179). This is an account of tests carried out by the Office Technique pour l'Utilisation de l'Acier in Belgium. The chemical compositions of the basic-Bessemer and open-hearth carbon steels, copper steels, chromium–copper steels and pure irons used are given. The plates were exposed to the atmosphere and immersed in sea and river waters at various stations. Examinations and tests were made at intervals up to five years and the results are tabulated and discussed.—J. C. R.

Rough Protection of Iron Exposed under Cover. L. Séguenot. (Métaux, Corrosion, Usure, 1941, vol. 16, Sept., pp. 61–65). Iron plates were given a preliminary treatment of brushing, sand-blasting, or pickling in various solutions, and then phosphatized, coated with a film of anti-rust oil, or exposed without further treatment, in the open air, but protected against rain. Brief details are given of the rusting of the plates after one month and six months. The sand-blasted and pickled plates gave similar results but the brushed

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plates gave poor results. Phosphatizing was not effective for more than one month, but the oil film gave good protection up to six months.—J. C. R.

Corrosion Coupons and Pipe Life Predictions—Revision of 1947. W. R. Schneider. (Corrosion, 1947, vol. 3, May, pp. 209–220). A detailed explanation is given of a method for predicting the life of pipelines buried in soil from measurements of the depths of pits formed on steel test-plates exposed to corrosion for one year at selected spots near the pipes.—C. O.

Anaerobic Corrosion of Iron in Soil. R. L. Starkey and K. M. Wight. (Corrosion, 1947, vol. 3, May, pp. 227–232). A general account is given of the corrosion of iron and steel brought about by sulphate-reducing bacteria.—C. O.

Preventing Corrosion in Gas-Condensate Wells. P. L. Menaul and P. P. Spafford. (American Institute of Mining and Metallurgical Engineers, Technical Publication 2229 : Petroleum Technology, 1947, vol. 10, July). The corrosion in gas-condensate or "distillate" wells has been shown to be caused by the formation of aqueous solutions of organic acids as the gases cool in ascending the well tubing. Laboratory and field tests indicate that the daily injection of as little as one quart of ammonium hydroxide into the casing annulus is effective in preventing this type of corrosion.—C. O.

Corrosion of Refinery Equipment—A Review. E. E. Kerns. (American Society of Mechanical Engineers : Corrosion, 1947, vol. 3, June, pp. 291–294). A brief account is given of the principal types of corrosion and protection methods met with in the oil-refining industry—C. O.

Mechanical and Metallurgical Control of Sulphuric Acid Corrosion in Petroleum Processes. E. R. Wilkinson. (American Society of Mechanical Engineers : Corrosion, 1947, vol. 3, May, pp. 252–262). The basic factors which influence the corrosion by sulphuric acid of oil-refining equipment are reviewed, and the methods which may be used to reduce this corrosion are discussed.—C. O.

Galvanic Aluminium Anodes for Cathodic Protection. R. B. Hoxeng, E. D. Verink, and R. H. Brown. (Corrosion, 1947, vol. 3, June, pp. 263–274). The development and testing of aluminium-alloy anodes for cathodic protection are described.—C. O.

Corrosion and Preventive Methods in the Katy Field. R. C. Buchan. (American Petroleum Institute : Corrosion, 1947, vol. 3, June, pp. 275–290). Typical examples of the types of corrosion of pipes and pumping equipment encountered in gas-condensate wells in Katy Field, Texas, are illustrated. Plastic coatings and neutralizing agents, such as soda-ash, appear to be useful in reducing corrosion, but chemical inhibitors have not been satisfactory.—C. O.

Laboratory Evaluation of Corrosion-Resistant Pigments and Vehicles. H. Zahn. (Paint, Oil, and Chemical Review, 1946, Sept. 5 : Corrosion, 1947, vol. 3, May, pp. 233–240). The use of potential/time curves for determining in the laboratory the protection afforded to steel by various corrosion-inhibiting paints and pigments is described.—C. O.

Notes on the Photography of Plates Used in Corrosion Tests. N. Leontovitch. (Métaux, Corrosion, Usure, 1941, vol. 16, Sept., pp. 65–66).—J. C. R.

The Study of Metals and Alloys by Means of the Metallographic Polarizing Microscope. Structure and New Optical Characteristics of Some Opaque Specimens. L. Capdecomme. (Métaux, Corrosion, Usure, 1941, vol. 16, Oct., pp. 77–87). An account is given of the field of application of polarized light in the study of metals and alloys and their corroded surfaces. A bibliography is appended.—J. C. R.

The Use of Wood Carbonisation Products in the Protection of Ferrous Metals. A. Lambriotte and J. Bompard. (Métaux, Corrosion, Usure, 1946, vol. 16, Nov., pp. 97–102). An account is given of the use of methylene oils in pickling solutions and in corrosion-resisting pastes and oils.—J. C. R.

ANALYSIS

Polarography in Germany. (British Intelligence Objectives Sub-Committee, 1947, Final Report No. 1458 : H.M. Stationery Office). Three types of polarograph and their accessories which were made in Germany during the war are described and details are given of methods used for the analysis of slags, ores, non-ferrous metals, and steels.—C. O.

Continental Wartime Developments in Spectroscopic Technique. E. H. S. van Someren. (Journal of Scientific Instruments, 1947, vol. 24, Sept., pp. 225–230). Wartime advances in the design and construction of spectrographs and light sources, and in spectroscopic techniques and methods, which have been reported in the Continental literature, are reviewed.—C. O.

Direct-Reading Device Provides Rapid Steel Analysis. E. R. Vance. (Steel, 1947, vol. 121, Sept. 22, pp. 92–94). A short description is given of a direct-reading spectrographic-analysis instrument which enables the manganese, silicon, chromium, nickel, molybdenum, copper, tungsten, and vanadium contents of steel to be obtained within 40 sec. of placing a sample in the electrode holder. The conventional photographic plate is replaced by a photomultiplier tube, the current from which charges a condenser. The time of discharge of the condenser is a measure of the content in the steel of the element being estimated.—C. O.

Qualitative Spectrographic Analysis and Its Practical Applications. J. Schoofs. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 171–179). The author describes the method and the advantages and disadvantages of spectrographic analysis, and reviews a number of practical problems that he has been able to solve in an industrial laboratory.—R. F. F.

Quantitative Spectrographic Analysis. P. Coheur. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 179–186). The principles of methods and instruments used in spectrum analysis are briefly described. Some practical examples of its use in metallurgy are given.—R. F. F.

The Spectrographic Analysis of a Complex Ferrous Alloy. C. H. R. Gentry and G. P. Mitchell. (Journal of the Society of Chemical Industry, 1947, vol. 66, July, pp. 226–232). A detailed description is given of a "concentration ratio" method for the spectrographic analysis of Ticonal alloy (6–9% Al, 12–18% Ni, 0–2·5% Ti, 2–5% Cu, 20–30% Co, 0–0·5% Si, remainder Fe). —C. O.

Determination of Chromium in Steels. M. J. Godart. (Revue Universelle des Mines, 1946, Series 9, vol. 2, No. 3, pp. 138–140). The photometric measurement of the width of spectral lines is applied to the quantitative determination of chromium in steels.—R. F. F.

Choosing a Spectrograph. P. Coheur. (Revue Universelle des Mines, 1946, Series 9, vol. 2, No. 1, pp. 25–26). The selection of a spectrograph is related to the type of analysis to be carried out, with special reference to the dispersion of the spectrograph.—R. F. F.

Some Comments on Possible Light Sources for Spectrochemistry. P. Swings. Revue Universelle des Mines, 1947, Series 9, vol. 3, No. 9, pp. 339–341). The use of special light sources, such as micro-spark, spark in vacuum, and the Schüler hollow cathode. is suggested for industrial spectroscopy.—R. F. F.

The Application of the Spekker Photoelectric Absorptiometer to the Analysis of Chromium Plating Solutions. W. J. Bayley. (Journal of the Electrodepositors' Technical Society, 1947, vol. 22, pp. 121–128). Spekker absorptiometer methods have been applied to the analysis of chromium-plating solutions. Trivalent and hexavalent chromium is determined directly, using orange and blue glass filters, respectively, while iron is estimated by means of the potassium thiocyanate complex, using a green filter.—C. O.

The Design and Use of a Photoelectric Absorptiometer for the Analysis of Solutions Employed for the Electro-Deposition of Hard Chromium. H. E. Styles. (Journal of the Electrodepositors' Technical Society, 1947, vol. 22, pp. 129–154). The control of the composition of chromic acid solutions used for electrodeposition is discussed. A Spekker type of photo-electric absorptiometer has been modified, in particular to make it suitable for turbidimetric measurements, and applied to the analysis of solutions for sulphate, chromic acid, iron, and trivalent chromium.—C. O.

X-Ray Absorption Analysis of Alloy Steels. H. Krainer. (Berg- und Hüttenmännische Monatshefte der Montanistischen Hochschule in Leoben, 1947, vol. 92, Apr.–May, pp. 61–65). After briefly reviewing the principles of X-ray absorption analysis, a method of applying it is described in which time is saved by using an electron counter to measure the intensity of the X-rays. As an example, details are given of the determination of tungsten in steel.—R. A. R.

Micro-Analysis. R. Chandelle. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 128–129). Mineral and organic reagents for micro-analyses are discussed.—R. F. F.

Rapid Spot-Tests for Some Constituent Elements in Ferrous and Non-Ferrous Alloys. J. Birckel. (Annales de Chimie Analytique, 1945, vol. 27, Mar., pp. 49–55). The technique of qualitative analysis of small quantities of solution, using organic reagents, is described and a table of tests for various metals is given.—R. F. F.

Determination of Sulphur in Steels. E. Leclerc. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 130–131). The author describes the results obtained in the determination of sulphur in a number of steel specimens, carried out at four different laboratories, and comments on the methods used.—R. F. F.

Determination of Nickel and Chromium in Steels. G. Pirenne. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 3, p. 124). The author estimates the experimental error in the spectrographic determination of nickel and chromium in steels.—R. F. F.

Determination of Manganese in Steels. R. Chataigner. (Annales de Chimie Analytique, 1945, vol. 27, June, pp. 111–112; Aug., p. 154.) The author studies the reduction of potassium permanganate solutions and the catalytic effect of manganous salts on this reduction. He explains the theory and describes a method for determining different amounts of manganese in steels or alloys with or without chromium.—R. F. F.

The Analysis of Nickel–Cobalt–Iron Alloys Used in Glass-to-Metal Seals. R. C. Chirnside, H. J. Cluley, and P. M. C. Proffitt. (Analyst, 1947, vol. 72, Aug., pp. 351–359). The authors discuss the analysis of nickel–cobalt–iron alloys of the "Kovar" and "Fernico" types, indicating the experimental difficulties and outlining methods which have proved satisfactory.—C. O.

Example of Potentiometric Analysis: Determination of Vanadium in Ferrous Metals. M. Chateau. (Chimie Analytique, 1947, vol. 29, Aug., p. 176). The author describes an accurate method for the potentiometric determination of vanadium, which is applicable to the analysis of ferrous metals.—R. F. F.

Determination of Aluminium by o-Oxyquinoline in Copper-, Nickel- and Ferrous Alloys. G. Lemine and J. Miel. (Annales de Chimie Analytique, 1946, vol. 28, Aug., pp. 147–148). The authors describe a rapid and precise method applicable to most metallurgical products containing aluminium, by which the direct determination of aluminium in the presence of a number of metals is possible. The limitations and accuracy of this method are also indicated.—R. F. F.

The Formation of Tervalent Manganese in Strong Sulfuric Acid Solution. J. F. G. Hicks, jun., and E. Krockmalski. (Journal of the American Chemical Society, 1947, vol. 69, Aug., pp. 1970–1971). The source of the violet colour in the solution when manganese dioxide is fumed with concentrated sulphuric acid has been determined experimentally. It is the manganic ion formed intermediately in the reaction $2MnO_2 + 4H^+ = 2Mn^{++} + 2H_2O + O_2$.—C. O.

A Comparison of the Abundance Ratios of the Isotopes of Terrestrial and of Meteoritic Iron. G. E. Valley and H. H. Anderson. (Journal of the American Chemical Society, 1947, vol. 69, Aug., pp. 1871–1875). An account is given of the measurement of the isotopic compositions of samples of purified terrestial and meteoritic iron by means of the mass spectrograph. No certain differences were found between individual samples or the two varieties.—C. O.

Methods of Determining Niobium and Tantalum in Basic Bessemer Slags. H. Barber. (Berg- und Hüttenmännische Monatshefte der Montanistischen Hochschule in Leoben, 1947, vol. 92, June, pp. 114–115)

Modern Methods of Analysis of Coal for Industrial Uses. H. De Rycker. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 131–135). The author comments on the results obtained and methods used with tests on coals, carried out by eight different laboratories.—R. F. F.

Yield of Volatile Matter of Coals. P. Schepers and P. Colson. (Mémoires de l'Association des Ingénieurs Sortis de l'Ecole de Liége, 1943, No. 4, pp. 135–141). The causes of variations of results obtained in determinations of the volatile matter in coals are discussed.—R. F. F.

Chemical Control of Bentonites Using the Color Test for Montmorillonite. A. Esme. (Annales de Chimie Analytique, 1946, vol. 28, Feb., p. 31). A simple chemical test for bentonite, using the blue colour produced by the reaction of montmorillonite with benzidine is described.—R. F. F.

HISTORICAL

Oriental "Damask" and the Damascus Swords. N. T. Belaiew. (Métaux et Civilisations, 1945, vol. 1, No. 1, pp. 10–16). The author reviews the history of Damascus steel and its introduction into Russia. He describes the manufacture and qualities of this steel and gives a short bibliography.—R. F. F.

The Metal Workers of the Sahara—Part I—With the Tuareg. R. Pottier. (Métaux et Civilisations, 1945, vol. 1, No. 2, pp. 31–40). The technique used by the craftsmen of the Tuareg for the working of iron, silver, and copper is described.—R. F. F.

The Laboratory Treatment of Archaeological Specimens of Iron. E. Salin. (Métaux et Civilisations, 1945, vol. 1, No. 3, pp. 49–61). The restoration of archaeological specimens of iron is described.—R. F. F.

Use of Metals in Susa. R. de Mecquenem. (Métaux et Civilisations, 1946, vol. 1, No. 4, pp. 77–88). In this

account of the various metals found in the remains of Susa, there is a brief reference to iron implements.—R.F.F.

Historical Notes on Metallurgical Technique. Part I. Progress in the Middle Ages—The Forge and the Blast-Furnace. B. Gille. (Métaux et Civilisations, 1946, vol. 1, No. 4, pp. 89–94). The progress made by the iron industry during the Middle Ages is described with special reference to the tilt-hammer and blast-furnace which date respectively from the 11–12th and 15th centuries.—R. F. F.

Historical Notes on Metallurgical Technique. Part 2. The Use of Coal in the Iron and Steel Industry. B. Gille. (Métaux et Civilisations, 1946, vol. 1, No. 5, pp. 99–104). The author reviews the development of the use of coal in the metallurgical industries from the 12th century onwards. A bibliography is included.—R. F. F.

ECONOMICS AND STATISTICS

The World Production of Iron. R. Durrer. (Von Roll Werkzeitung, 1946, vol. 17, Sept., pp. 293–294). Statistics on the world production of iron since 1870 and the proportion used for steelmaking are presented and discussed.—R. A. R.

The War-time Position of the British Iron and Steel Industry and Its Modernization Projects. (Génie Civil, 1947, vol. 124, Oct. 15, p. 400). The present position and projected modernization of the British iron and steel industry are outlined.—R. F. F.

Australian Iron and Steel Industry. (British Steelmaker, 1947, vol. 13, Sept., pp. 476–477). In a brief review of the Australian iron and steel industry it is stated that the iron-ore resources of the Commonwealth are equal to at least 100 years' supply. The industry has developed greatly since 1935, and ingot production has more than doubled in that time.—J. R.

Steel on the Pacific Coast. L. H. Duschak. (Chemical and Engineering News, 1947, vol. 25, Sept. 22, pp. 2707–2709). Recent developments in steel production on the west coast of the United States are described. Production was stimulated by wartime requirements, but local supplies of raw materials are generally small. Comparative figures are given showing the outputs of cast iron and steel in 1938 and 1945.—M. A. V.

Industrial Planning. Z. Majewski. (Hutnik, 1946, vol. 13, May, pp. 270–271). [In Polish]. The planned production in Poland of various products including ores, coal, iron, and non-ferrous metals for the years 1946–1948 is compared with the output in 1937–1938. —W. J. W.

MISCELLANEOUS

Developments in Metallurgy at Birmingham University. L. Aitchison. (Journal of the Birmingham Metallurgical Society, 1947, vol. 27, June, pp. 281–295). The aims of the new Department of Industrial Metallurgy at Birmingham University are discussed, and an outline is given of the course of instruction and the laboratory facilities available.—C. O.

Importance of Metallurgy for Special Sections of Engineering Industry. F. Drastik. (Hutnické Listy, 1947, vol. 1, No. 10, pp. 217–219). [In Czech].

Radioactive Tracers in Metallurgical Research. E. S. Kopecki. (Iron Age, 1947, vol. 160, Sept. 4, pp. 60–64). A research programme being carried out at the Carnegie Institute of Technology is outlined. The rates and mechanisms of metallurgical reactions in solids and liquids are being investigated with the aid of radioactive tracers.—C. O.

Research in the Steel Industry. G. Grenier. (Revue de l'Industrie Minérale, Comptes Rendus, 1946, July, pp. 110–115). The iron and steel research organizations of France, Sweden, Germany, and Great Britain are described, with notes on the more important researches carried out by these organizations.—R. F. F.

The Training of Metallurgists for the Steel Industry. J. Chipman. (American Iron and Steel Institute, May, 1947, Advance Copy). The nature of metallurgy requires that its students should be trained both in the physical sciences and engineering. To attempt to encompass these subjects in the normal four-year university course, without neglecting the cultural aspects of education is a task of great magnitude. The author considers that the time available would be more profitably employed in obtaining a thorough grounding in basic principles and the fundamental sciences than in learning details of specialized industrial processes.—C. O.

Telemetering Systems: Remote Indication of Quantities. (*The Times* Review of Industry, 1947, Sept. 22). This article describes systems for the remote indication of quantities such as temperature, gas and steam flows, etc., by electrical means, over wire systems and over Post Office lines. In the latter case, photo-electric cells and appropriately tuned circuits produce indications unaffected by variations in the electrical constants of the lines—J. P. S.

New Nozzle Designs for Low and Medium Reynolds Numbers. W. Koennecke. (Iron and Steel Institute, 1947, Translation Series, No. 320). This is a translation of a paper which originally appeared in Forschung auf dem Gebiete des Ingenieurwesens, 1938, vol. 9, May–June, pp. 109–125. The author gives details of a large number of tests made to determine a design of nozzle which would have a constant value of the Reynolds number in the range below 10^4 and which could be used with a variety of aperture ratios. The most suitable design is shown to be a combination of a continuously curved inlet profile (a quarter-circle) with a cylindrical outlet.—C. O.

The Design of Automatic and Manually Operated Control Systems. A. Porter. (Transactions of the Society of Instrument Technology, 1947, vol. 1, Jan., pp. 22–40). A detailed account is given of the theory and design of those types of automatic and manually operated control systems which are based on position.—C. O.

Mechanism of Countercurrent Gas Liquid Flow through Packed Towers. F. A. Zeny. (Chemical Engineering Progress, 1947, vol. 43, Aug., pp. 415–428). A theory of countercurrent flow through packed towers is derived from analogies to flow through valves and orifices. "Flooding point" is defined according to the theory, and an equation is obtained which may be used to calculate the pressure drop at various flow rates.—C. O.

Base-Exchange or Lime-Soda Water Softening? F. C. Blight. (Mechanical World, 1947, vol. 122, Aug. 22, pp. 179–182). The mechanical, chemical, and economic factors involved in the choice of the type of water softener for boilers using less than 2000 gal./hr. of feed water and operating at pressures of less than 200 lb./sq. in. are considered.—C. O.

Some Practical Results Obtained with the Present Methods of Treating Feed Water. R. Vinçotte. (Mémoires de l'Association des Ingénieurs sortis de l'Ecole de Liége 1943, No. 2, pp. 81–100). The author describes various ways of purifying feedwater as precautions against scaling and corrosion, and discusses the effects of impurities carried over by the steam.—R. F. F.

Use of Sodium Hexametaphosphate for Testing Condenser Water in Boiler Plants in the Ruhr. M. Chalmot. (Revue de l'Industrie Minérale, 1947, Aug.-Sept., pp. 515–520).—R. F. F.

Research in the Iron and Steel Industry of Great Britain. (Iron and Coal Trades Review, 1947, vol. 155, June 20, pp. 1151–1165). In this, the first of a series of articles in which the research facilities available to the iron and steel industry in Great Britain will be outlined, the scope of the research work which has been and is being carried out in the works' laboratories is illustrated by descriptions of the activities at various individual establishments. References to 71 papers dealing with the results of industrial research are appended.—J. R.

Research in the Iron and Steel Industry of Great Britain. (Iron and Coal Trades Review, 1947, vol. 155, Aug. 15, pp. 299–310). This is the second of a series of articles reviewing research work in the British iron and steel industry. The organization of co-operative research is dealt with, and an outline is given of the activities of the Iron and Steel Institute, the Iron and Steel Industrial Research Council, the British Cast Iron Research Association, and the various Universities and Colleges. The development of the British Iron and Steel Research Association is discussed, and its internal organization and the scope of its work are described in detail.—J. R.

High- and Low-Pressure Processes in Oxygen Manufacture. S. Ya. Gersh. (Kislorod, 1947, vol. 4, March–April, pp. 1–6). [In Russian]. The high first cost of high-pressure plants for manufacturing liquid oxygen is easily offset by low power consumption, compared with low-pressure plants, and the latter should be used only if circumstances make it essential. Cheap gaseous oxygen can be made at various pressures according to the size of the plant. The Linde, Claude, and Clerk processes require small and medium plants. The Linde-Frenkl is a large plant involving constructional difficulties, but it is considered the most economical process.—R. A. H.

Linde-Fränkl Process for Oxygen Separation. (Coke and Gas, 1947, vol. 9, Aug., pp. 236–240). An outline is given of the operation of the Linde-Fränkl process for the production of oxygen. An attempt is made to estimate the cost of production in terms of current British conditions, on the basis of data obtained from various German works using the process. Assuming power to cost 0·36*d.* per kWh., the cost is calculated to be 17*d.* per 1000 cu. ft. of 98% oxygen.—C. O.

Electrolytic Conductivity as a Method for Studying Electronic Transitions in Elements—Applications to Iron, Nickel and Cobalt. W. R. Ham and C. H. Samans. (Transactions of American Society for Metals, 1947, vol. 39, pp. 73–99).—R. A. R.

Protective Packing for Shipment or Storage. J. E. Foley. jun. (Corrosion and Material Protection, 1947, vol. 4, Sept.-Oct., pp. 6–7). A short description is given of the "Cocoon" process for the protection of metal parts during shipment or storage. The article to be covered is surrounded by a grid of tape—not touching the metal. The grid is filled in with cobweb-like filaments from a spray gun, and the whole coating is reinforced by several layers of a plastic film also sprayed from a gun. —C. O.

Metallurgical Progress in the Valve, Fitting and Piping Industry. J. J. Kanter. (Metal Progress, 1947, vol. 52, Oct., pp. 629–631). A short survey is made of the developments in the valve and fitting industry since the introduction of superheated steam.—C. O.

Tracer Isotopes in Metallurgy. J. K. Stanley. (Nucleonics, 1947, vol. 1, Oct., pp. 70–77). An account is given of the principles and the history of the use of radioactive isotopes in metallurgical research. Problems of diffusion, oxidation and corrosion, metallography, wear, steelmaking, heat-treatment, and chemical analysis which might be solved with the aid of isotopes are indicated.—C. O.

BOOK NOTICES

BAEYERTZ, M. "*Nonmetallic Inclusions in Steel.*" 8vo, pp. 135. Illustrated. Cleveland, Ohio, 1947: American Society for Metals.

The purpose of this book is to bring together enough of the basic principles which govern the formation of inclusions, the estimation of their amount in commercial products, and the identification of their mineralogical character, to afford the beginner an introduction to the subject. The first chapter is introductory, the next five deal with the origin of inclusions in ingots, and these are followed by chapters on the formation of inclusions in subscales, the effects of rolling and forging, methods for the estimation of the type and amount of inclusions, and methods for their identification. An appendix is included which lists some inclusion minerals arranged in the approximate order of increasing hardness by Mohs' Scale.—R. E.

BRITISH WELDING RESEARCH ASSOCIATION. "*Symposium on Metallurgy of Steel Welding.*" La. 8vo, pp. 104. Illustrated. London, 1947: The Association. (Price 10*s.*).

This book contains all the papers and discussions presented at a series of meetings organized by the British Welding Research Association giving the results of modern metallurgical research in relation to arc welding. The contents of the book include some of the most comprehensive papers produced in recent years on welding science. Notable among these are "Constitution of Weld Metal," by W. Andrews, "A Preliminary Investigation of the Constitution of Mild Steel Arc-Weld Deposits," by H. A. Sloman, T. E. Rooney and T. H. Schofield (*see also* Journ. I. and S.I. 1945, No. II, pp. 127P–153P), "An Analytical Examination of Weld Deposits from Commercial Mild-Steel Electrodes to Specification B.S.S. 639A, and the Manufacture of Artificial Weld Metal," by G. L. Hopkins, and "Comments on the Rôle of Hydrogen in Relation to the Cracking of Alloy Steels on Welding," by G. L. Hopkin.

Important contributions to the Symposium were made by Dr. L. Reeve, whose papers included "The Relation between the Hydrogen Content of Weld Metal and Its Oxygen Content," "Influence of Sulphur and Phosphorus on Weldability of Mild Steel," and a note on "Cracking of Welded Gas Mains." Drs. Wheeler and Kondic submitted a paper on "Effect of Initial Heating Temperature on the Mechanical Properties of an Air-Hardened Ni–Cr–Mo

Steel" giving an account of investigations at Birmingham University and presenting for the first time data on transformation products formed after quenching from abnormally high temperatures.

A paper by P. K. Gledhill, entitled "An Apparatus for the Determination of the Viscosity of Welding Slags" gives the results of preliminary investigations. An important contribution is the paper by Dr. L. H. Orton, entitled "Electro-Physics of the Welding Arc."—R. E.

GRAVES, H. B., jun. "*The Mineral Key.*" First Edition. Sm. 8vo, pp. viii + 178. New York, 1947: McGraw-Hill Book Co., Inc.; London: McGraw-Hill Publishing Co., Ltd. (Price $4.00).

This small book is designed to enable the amateur mineralogist to identify and become better acquainted with his mineral specimens without the use of expensive equipment or special training in mineralogy. Over 100 pages are devoted to tables giving the names of the various minerals, their chemical formulæ, composition, and distinguishing characteristics.—R. E.

JUDGE, A. W. "*Modern Gas Turbines.*" 8vo, pp. xi + 311. Illustrated. London, 1947: Chapman and Hall, Ltd. (Price 28*s.*).

This book describes briefly the history and development of various types of gas turbines and then deals with questions of comparison, thermodynamics, and efficiency as between gas and steam turbines, diesel engines, etc. Later chapters discuss closed-cycle gas turbines, exhaust-gas turbines for supercharging engines, aircraft application, performance and the conditions governing the selection of turbine materials. A comprehensive bibliography is included of references to works mentioned in the text and others which should be consulted for further study.—R. E.

OLLARD, E. A., and E. B. SMITH. "*Handbook of Industrial Electroplating.*" 8vo, pp. 287. Illustrated. London, 1947: Metal Industry. (Price 15*s.*).

In this book the authors have attempted to collate under one cover, by drawing largely on current literature, as much data as possible relating to the electroplating industry. It is in no sense a textbook—all theory has been omitted; it describes merely the practical part of the installation and maintenance of electrodeposition plant. The book is divided into the following eight parts: Electrical equipment; deposition plant; suitable formulæ; special formulæ; testing depositing solutions; testing deposits; glossary; miscellaneous, and tables.—R. E.

OLLIVER, C. W. "*The Intelligent Use of the Microscope.*" 8vo, pp. vii + 182. Illustrated. London, 1947: Chapman and Hall, Ltd. (Price 12*s.* 6*d.*).

This book is intended to be a guide for the student to an understanding of the possibilities and limitations of the microscope, the principles of its design, and the precise functions of its components and acccessories. Chapters are given on optical theory, numerical aperture and resolution, the components and accessories of the microscope, the selection of the right equipment, illumination, the microscope in use, measurements and counts, filters, photomicrography, and special instruments and applications. A bibliography of the literature on the subject is included. —R. .E.

NEW PUBLICATIONS

AMERICAN SOCIETY FOR TESTING MATERIALS. "*Symposium on Testing of Bearings.*" 8vo, pp. 65. Philadelphia: The Society. (Price $1.50).

AMERICAN SOCIETY FOR TESTING MATERIALS. "*Symposium on Atmospheric Weathering of Corrosion-Resistant Steels.*" 8vo, pp. 85. Philadelphia, 1946. (Price $1.50).

BAEYERTZ, M. "*Nonmetallic Inclusions in Steel.*" 8vo, pp. 135. Illustrated. Cleveland, Ohio, 1947: American Society for Metals.

BRADY, G. S. "*Materials Handbook.*" Sixth Edition. 8vo, pp. 831. New York, 1947: McGraw Hill Book Co., Inc.; London: McGraw-Hill Publishing Co., Ltd. (Price 35*s.*).

"*British Fuel and Power Industries.*" London: Political and Economic Planning. (Price 30*s.*).

BRITISH RHEOLOGISTS CLUB. "*Essays in Rheology.*" A Contribution to the Literature of the New Science, based on the 1944 Oxford Conference of the British Rheologists' Club. 8vo, pp. viii + 103. Illustrated. London, 1947: Sir Isaac Pitman and Sons, Ltd. (Price 12*s.* 6*d.*).

BRITISH WELDING RESEARCH ASSOCIATION. "*Symposium on Metallurgy of Steel Welding.*" La. 8vo, pp. 104. Illustrated. London: The Association. (Price 10*s.*).

COMMISSION TECHNIQUE DES ETATS ET PROPRIÉTÉS DE SURFACE DES MÉTAUX. "*Journées des Etats de Surface, Paris, 23–26 octobre, 1945.*" 4to, pp. viii + 273. Illustrated. Paris, 1946: Office professionnel général de la Transformation des Métaux. (Price 1050 francs).

LARKIN, E. J. "*Works Organisation and Management.*" Second Edition. London: Sir Isaac Pitman and Sons, Ltd. (Price 40*s.*).

SOCIETY OF AUTOMOTIVE ENGINEERS. "*SAE Handbook.*" 1947 Edition. 8vo, pp. 822. New York: The Society. (Price $10.00).

SPECTROGRAPHIC DISCUSSION GROUP, GLASGOW. "*Suggested Definitions of Terms Used in Spectrographic Analyses.*" 8vo, pp. 11. Aberdeen, 1946: The University Press.

SPRING, H. M., jun. "*Pressure Vessels for Industry.*" New York: McGraw-Hill Book Co., Inc., London; McGraw-Hill Publishing Co., Ltd. (Price 17*s.* 6*d.*).

STAUB, M. "*Recherches sur la Corrosion des Métaux par le chloroforme et le tétrachlorure de carbone.*" (Thèse d'ingénieur-docteur). Pp. 55. Illustrated. Paris, 1946: Université de Paris.

VEREIN DEUTSCHER EISENHÜTTENLEUTE. "*Anhaltszahlen für die Wärmewirtschaft insbesondere auf Eisenhüttenwerken.*" Nach Zahlenunterlagen und Versuchsergebnissen, gesammelt vom Verein deutscher Eisenhüttenleute durch dessen Energie- und Betriebswirtschaftsstelle. Bearbeitet, ergänzt und herausgegeben von Kurt Rummel. 4. Auflage der "Anhaltszahlen für den Energieverbrauch in Eisenhüttenwerken." La. 8vo, pp. viii + 224. Illustrated. Düsseldorf, 1947: Verlag Stahleisen, m.b.H.

WERNER, E. "*Metallische Uberzüge auf Elektrolytischem und Chemischem Wege und das Färben der Metalle.*" Third Edition, pp. 187. Munich, 1946: Carl Hanser-Verlag.

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